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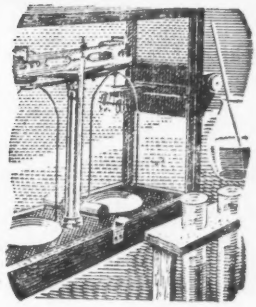

Relationships between Long- and Short-Time Creep and Tensile Properties of a Commercial Alloy

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SYNOPSIS

Experimental results are given for short-time creep tests, two forms of tension test, and creep-recovery tests, on Nimonic 80, and comparisons are made between these and with long-time creep data for Nimonic 80 type alloys. The results were used to guide the development of a theory of deformation, based upon the Nutting equation and the Boltzmann memory principle, which meets a variety of requirements. Satisfactory relationships are established amongst the various results and the long-time data.

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AN APPRAISEMENT OF THE PROBLEM OF MECHANICAL PROPERTIES

One of the major objectives of the study of metals is to produce materials having adequate mechanical properties, and one of its major handicaps is that no simple comprehensive measure of these properties has been found. The mechanical behaviour of a metal in different machines and structures is apparently not dependent upon a single property, and its behaviour in different circumstances is consequently estimated by the engineer from the results of empirical tests that approximate to a greater or lesser degree to the various conditions of use. Behaviour is affected by small changes of composition and treatment, and such changes in general necessitate fresh measurements. This laborious procedure is held to be inevitable in view of the complex structure of metals as revealed by the microscope and X-rays, and by the observation that each constituent of the structure may change in a variety of ways, according to the loading, temperature, amount of deformation, and (whether the material is deformed or not) the time.

Yet the results of the more fundamental type of mechanical test by no means reflect the complexity and diversity of the metallurgical scene. For example, although different metals exhibit different forms of creep curve, physicists and engineers have thought it worthwhile to seek a general mathematical expression for creep. Much controversy has existed between those with rival suggestions and theories to defend upon the precise expression to use; however, in view of the background complexity, the differences between curves seem to be of much less significance to a general understanding of properties than the fact that the curves of materials of such widely differing structure as metals, ceramics, pitch, plastics, and wood are so similar. It is not intended to suggest that differences are not significant, but rather that the general shape of the curves may perhaps be governed by some common principle, which is very much simpler than the aggregation of special detail that is presented when attention is confined to any single class of materials, particularly metals.

The reasons for presenting this paper to metallurgists are that the specific problem considered was one often faced by them, namely, that of predicting long-time creep from short-time tests, and that the material studied was the alloy Nimonic 80. It has, however, been necessary to go outside the normal

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field of metallurgical interest to find a satisfactory explanation of the results.

One complication is unavoidable. It is that, whereas the mechanical properties of gases concern the relation between pressure, volume, and temperature, those of metals are expressed not merely by the relation between the corresponding variables stress σ, strain ε, and temperature T, for time t occurs in addition. The fact has two consequences which are not always recognized. One is that the mere existence of four primary variables demands extensive experiments and a systematic and consecutive procedure for sorting out their relationship; the other is that the additional variable time, whose relevance is manifested by the fact that materials change with time whether they are loaded or not, presents a fresh and unfamiliar problem in science. A mechanical 'property' is not a property in the usual physical sense of the term.[1]

OUTLINE OF WORK

The work started in 1948 with a practical test of the possibility of predicting long-time creep from short-time results on the basis of the suggestion (due originally to Zener) that an increase of temperature has no other effect than to multiply the rate of extension by some numerical factor. Two quantitative expressions of this suggestion were tried; one, due to MacGregor and Fisher,[2] that temperature and straining rate $\dot{\varepsilon}$ occur only in such a combination as

$$T\left(1-k\log\frac{\dot{\epsilon}}{\dot{\epsilon}_0}\right)=T_m \quad \ldots\ldots\ldots\ldots\ldots (1)$$

and another, suggested by Graham[3] and later independently proposed and supported by Larson and Miller,[4] that temperature and time occur only in such a combination as

$$T\,(1+a\log t)=\theta \quad \ldots\ldots\ldots\ldots\ldots (2)$$

A given value of T_m or θ is reached in a shorter experimental time by raising the temperature.

For satisfactory extrapolation, it is not alone necessary to know the law of extrapolation, but a firm basis must be provided from which to extrapolate: *i.e.*, a satisfactory co-ordination of the shorter-time data. Any set of creep or tensile-test data expresses in principle a quantitative relationship between the four variables mentioned above; and co-ordination implies the fitting of a mathematical formula. Interest was accordingly taken in the various equations between them that have been proposed.[5–8] The equations are analogous to the law $pv = RT$ for gases and are termed 'equations of state.' However, the work of Los, reported by Orowan,[9] Dorn, Goldberg, and Tietz,[10] and Tietz and Dorn,[11] appears to have brought their development to a standstill, for it clearly showed that the four quantities could not in general be related by any form of equation in which any one quantity is uniquely determined when values of the remaining three are given. These objections had therefore to be met. Owing to the presence of four variables, the results of one kind of test alone were not sufficient to guide improvements.

A stage-by-stage method was followed,[12] beginning with a promising form of equation and progressively modifying it to meet the specific objections that were raised by the experimental results. The initial equation, which embodies equation (2), is

$$\frac{\sigma}{\sigma_s}=\epsilon^{\mu}\left(\frac{t_s}{t}\right)^{\zeta T} \quad \ldots\ldots\ldots\ldots\ldots (3)$$

in which σ_s, t_s, ζ, and μ are constants and ε is taken to be the non-elastic strain. It was derived in a previous paper[3] and there shown to provide a reasonably satisfactory representation of families of creep curves. The equation was modified by supposing that it

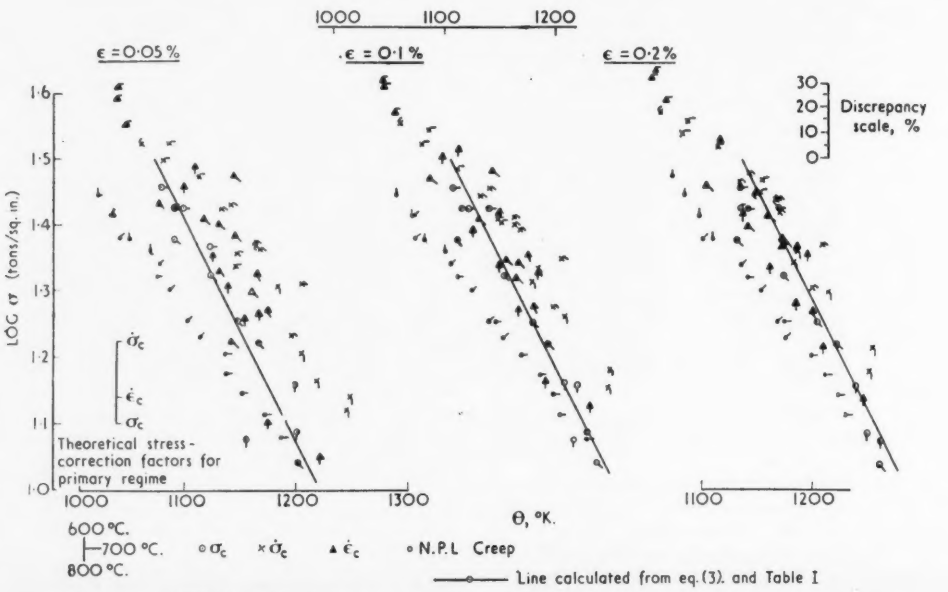


Fig. 1—Comparison of results on basis of θ (log $t_s = -17{\cdot}3$)

applies to incremental rather than finite changes, the increments being combined according to a principle due to Boltzmann; each of the many modes of deformation of a metal was supposed to obey an expression of the type found; evidence for the way temperature should enter into the argument led finally to a theory that substantially met not only the immediate requirements of the experiments but also certain essential requirements of a more general nature. The present paper gives the main stages of the developing argument and their experimental justification.

DETAILS

An interest in engineering materials made to commercial tolerances led to experiments on specimens cut from Nimonic 80 drawn at random from the workshop stores. This material was eventually shown to be considerably stronger, owing to the manufacturers' improvement of the alloy, than the original form of the material, and it was then found that no extensive long-time data referring explicitly to the material used could be provided. The long-time data with which comparisons were finally made were, accordingly:

(i) Those for Nimonic 80 tabulated for times up to 10,000 hr. and issued by the Mond Nickel Co., Ltd.[13]

(ii) Some creep curves extending up to 1300 hr. for an early heat of Nimonic 80 and made available to the authors by courtesy of the National Physical Laboratory

(iii) Some creep curves, which extended up to 22,000 hr. and which related to an early heat (Z88) of Nimonic 80A, kindly supplied by Mond Nickel as being, of the data available, that likely to correspond most closely to the experimental material.

Most of the short-time tests were completed within 24 hr; the remainder ranged up to 74 hr., with three creep tests extending respectively up to 134, 200, and 250 hr. The temperature range was 700–850° C., *i.e.*, 50–100° C. above that for normal use of the material.

Three main forms of loading in tension were used: constant load, or, effectively, constant stress (σ_c), as in the usual creep test; constant rate of plastic straining ($\dot{\epsilon}_c$); and constant rate of application of stress ($\dot{\sigma}_c$). The two latter are two forms of tensile test. A few measurements were also made of the recovery of strain after removal of load.

Experimental details call for little comment. Owing to the smallness of the strains of practical interest, no specific arrangement to maintain constant stress during extension was necessary, but small corrections were made to the results at the larger strains. The specimens were of 2 in. gauge length and 0·253 in. dia., and were taken from casts AT.51, BE.4, BD.76, BH.96, CK.93, and DP.5. All material received the standard treatment of 8 hr. at 1080° C. followed by 16 hr. at 700° C., and most specimens had, for reasons that were later abandoned because no systematic effects whatever could be detected, various short periods at 650° and 700° C. Load and extension were recorded photographically; the control of temperature and accuracy of measurement of all quantities were well within (except for strains of the order 10^{-4}) the error permitted by the random variations in the material. Variations of the order found between repeat tests on similar samples of the material are quite normal for a complex alloy.

PRELIMINARY WORK THAT SET A THEORETICAL PROBLEM

Like many others that have been suggested, equation (3) is a particular form of an equation due to Nutting[14, 15, 16]:

$$\psi \epsilon = \sigma^{\beta} t^{\kappa} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (4)$$

in which ψ, β, and κ are constants at a fixed temperature. In view of the remarks in the first section

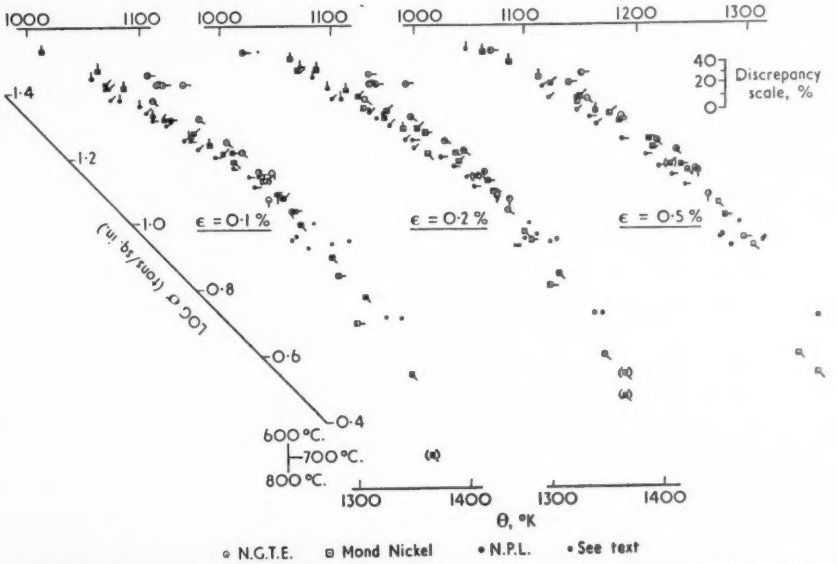


Fig. 2—Comparison of Mond Nickel, N.P.L., and N.G.T.E. creep results (log t_s = −17·3)

Table I

EXPERIMENTAL VALUES OF CONSTANTS IN EQUATION (6)

N.G.T.E. Material

	Decelerating Creep		Accelerating Creep
Log σ_s	6·35		5·67
Log t_s	(−16·26)	−17·4	(−18·43)
μ	0·39		0·14
ζ	$1{\cdot}92 \times 10^{-4}$		

σ in tons/sq. in., time in minutes

the reason for equation (4) being of especial interest is that, with appropriate values of the constants, it represents the behaviour of either the ideal perfectly elastic material or of the ideal perfectly viscous material, or, with intermediate values of the exponents, as has been experimentally demonstrated,[1,6,7] that of real materials of a diversity of classes loaded in different ways which are neither perfectly elastic nor perfectly viscous, but which exhibit some intermediate combination of these two extremes. The equation is an ' equation of state,' and must therefore for the present be regarded as a useful tool rather than as an exact and final statement of a physical law.

It becomes identical with equation (3) when

$$\beta = 1/\mu,\ \kappa = \zeta T/\mu,\ \psi = \sigma_s^{\,1/\mu}\, t_s^{\,\zeta T/\mu} \quad \ldots\ldots\ldots(5)$$

If t and T are combined in the variable θ defined in equation (2),

$$\frac{\sigma}{\sigma_s} = \epsilon^{\mu}\, t_s^{\,\zeta\theta} \quad \ldots\ldots\ldots(6)$$

and only three variables, σ, ε and θ, now appear. The relation between t_s and a in equation (2) is

$$-a = 1/\log t_s.$$

A formula very similar to equation (6) may be derived[3] with T_m in place of θ. Indeed, the only difference between equations (1) and (2) in the present context is that they represent different temperature variations of the quantities β, κ, and ψ. The variation over the total range of temperature covered (200° C.) is, however, well within the scatter of the data analysed; no further reference to equation (1) will accordingly be made.

A convenient way of testing this set of principles is to cross-plot a group of results in the form (*cf.* equation (6)) log σ *vs.* θ for various constant values of ε. The points should fall on a straight line. This has been done in Fig. 1, which shows, for three typical strains, the $\dot{\varepsilon}_c$, $\dot{\sigma}_c$, and σ_c short-time results together with the longer-time N.P.L. creep results. Figure 2 contains creep results alone, namely, those tabulated by Mond Nickel, together with the N.P.L. and σ_c results from Fig. 1, with also some further σ_c results, distinguished by blocked-in squares, which extend the range of stress and were obtained at a later stage. Temperatures are shown, as indicated by the legends, by tabs on the points; times are in minutes and stresses in tons per sq. in. The greater scatter of the $\dot{\varepsilon}_c$ results at all strains and of the $\dot{\sigma}_c$ and σ_c results at $\varepsilon = 0{\cdot}05\%$ is due to experimental causes, but appears to be unsystematic.

When the decelerating and accelerating stages of creep were considered independently, as discussed by Graham[3] and below, equation (6) could be fitted to any of the sets of data to within the scatter. The lines in Fig. 1 relate to the σ_c circles for which the appropriate constants are given in Table I. Their slope is ζ, and the same value of ζ was found to apply to both stages of creep. The lines are typical of the fit that can be obtained by using equation (6) for the other methods of loading. A large number of diagrams similar to Fig. 2 relating to long-time creep have been published by Larson and Miller.[4]

The rather large scatter of the points does not conceal a number of significant facts. If the blocked-in points (to be considered later) are excluded, no justification is found for using different values of log t_s for different sets of data, and −17·3 with time in minutes is suitable for them all. In Larson and Miller's units (hr.) this becomes of magnitude 19·1, in reasonable agreement with their 20.

Equation (6), like any other equation between σ, ε, and θ, states that, for given values of σ and ε, there is only one value of θ. Since such an equation draws no distinction between different methods of loading, the points for all methods should fall upon a single line. But Fig. 1 clearly shows the $\dot{\sigma}_c$ points to lie above the σ_c points, and the more scattered $\dot{\varepsilon}_c$ points to lie generally between the two. The general objection mentioned in the ' Outline of Work ' to an equation of state is therefore of practical importance.

Nevertheless, the points for the different short-time loading histories appear to be related, because a parallel displacement of any one set would bring it into reasonable agreement with any other.

In Fig. 2, the Mond data exhibit an abrupt change of slope in the neighbourhood of log $\sigma = 1{\cdot}1$, a well-known feature which is found in an even more marked form in all but one of Larson and Miller's curves. Such discontinuities, the possible occurrence of which makes any direct extrapolation of creep results extremely unreliable, are associated with changes in the nature of the flow[17,18] and are often attributed to changes in the material that may cause a sudden weakening after a long period of use. The important result shown by Fig. 2 is that it occurs also in the short-time σ_c tests and occurs, moreover, at values of log σ and θ which are substantially the same as in the long-time tests. The values are compared for log $t_s = -17{\cdot}3$ in Table II. The other entries in the Table are considered later.

Another cause of sudden weakening is the onset of tertiary creep. In the N.P.L. curves it occurred at strains between 0·12 and 0·14%, and at corresponding stresses in the short-time σ_c experiments over substantially the same range (0·11–0·17%*). The Mond tabulated data do not identify the point of onset.

The encouraging conclusion is that there is evidently some quantitative connection between results for different methods of loading and for long- and short-time tests, and also that some features of this connection are exhibited by equations (6) and (2). However, three specific and distinct problems require attention. One is that of the change to equation (6) needed to make it account for the different levels in

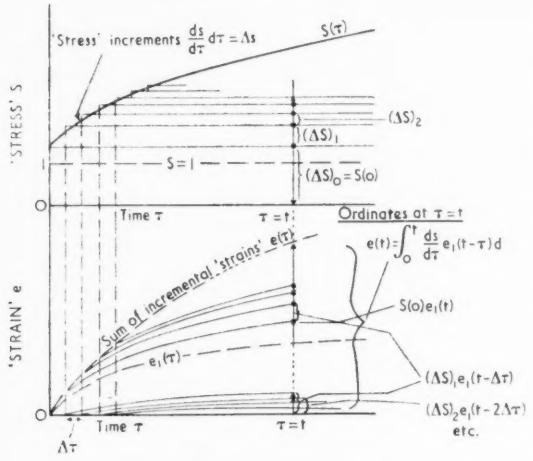
* These figures refer to log-strain/log-time curves, the onset being taken at the point at which the experimental curve departs from the straight line through the lower-strain points by the amount 10^{-4} in strain.

Fig. 1 of the $\dot{\sigma}_e$, $\dot{\varepsilon}_e$, and σ_e results. Another is that of the method of dealing, in cross-plots like Fig. 2 and Larson and Miller's diagrams, with the different linear sections separated by abrupt transitions, and a third is to provide a more rational basis for the apparently satisfactory use, shown by Graham[3] and by the present work, of separate equations of type (6) to represent the accelerating and decelerating stages of creep.

THEORY OF EFFECT OF LOADING HISTORY

If the question of temperature is set aside for the moment, the first of these problems is that of improving the Nutting equation (4) whilst retaining its valuable features. The above results and those of Dorn and others[10,11] permit the assumption that, instead of equation (4), which is a relationship between the measured *finite* values of σ, ε, and t, an exactly similar equation should be used to relate the *incremental changes* $\delta\sigma$, $\delta\varepsilon$, and δt of the variables. This assumption can be applied in different ways. One that meets immediate requirements and introduces an essential new feature, whereby it becomes mathematically possible to account for a change of the material with time and loading, is to combine equation (4) with the Boltzmann memory principle.

This principle was originally proposed to account for observations on creep recovery, and it has been experimentally justified in a variety of circumstances involving both small and large strains and a variety of materials (*e.g.*, Leaderman,[19] who gives earlier references, Taylor,[20] and Henderson[21]). The ' strain ' e for any finite ' stress ' s is considered to be the superposition of the incremental ' strains ' that develop from each increment of the ' stress,' each from the particular moment when the particular increment is applied. Subsequent ' stress ' increments have no influence upon the effects initiated by previous increments, the various effects being merely additive.



$$e(t) = \int_0^t e_1(t-\tau)\,\frac{d}{d\tau}\left[s(\tau)\right]d\tau$$

Fig. 3—Boltzmann's superposition principle

Table II

CO-ORDINATES OF MAIN θ TRANSITION

Creep Strain	Log t_s (min.)	Mond Data		N.G.T.E. Data	
		log σ (tons/sq. in.)	θ, ° K.	log σ (tons/sq. in.)	θ, ° K.
0·1%	−17·3	1·15	1205	1·16	1207
	−24·8	1·04	1170	1·13	1173
0·2%	−17·3	1·17	1232	1·17	1234
	−24·8	1·13	1170	1·13	1193
0·5%	−17·3	1·14	1259	1·19	1246
	−24·8	1·12	1186	1·19	1186
Mean for all strains	−17·3	1·15	1232	1·17	1229
	−24·8	1·10	1175	1·14	1184

The principle is illustrated in Fig. 3, in which it is to be supposed that unit ' stress ' applied abruptly at time $\tau = 0$ and maintained constant thereafter produces a ' strain ' e_1 which will be written as $e_1(\tau)$ to indicate its change with time. Then a varying ' stress ' written as $s(\tau)$ is regarded as a succession of abrupt increments $d/d\tau\ [s(\tau)]d\tau$, each of which initiates a corresponding proportionate increment of ' strain.' At any particular time $\tau = t$, the resulting ' strain ' $e(t)$ will be the sum of all the incremental ' strains ' initiated since the beginning of loading, *i.e.* (*see* Fig. 3)

$$e(t) = \int_0^t e_1(t-\tau)\,\frac{d}{d\tau}\left[s(\tau)\right]d\tau \quad \ldots\ldots\ldots\ldots (7)$$

In the original use of the principle, e and s were taken to be the actual strain and stress, thereby introducing the assumption that strain is proportional to stress, as if the material were perfectly elastic. This limitation, however, which is unacceptable in the present problem, is unnecessary, for e and s may be taken to be any functions of strain and stress. The assumption involved in this more general form of the principle is that the quantities concerned obey a linear differential equation. There is no limit to the complexity of the equation, provided only that it is linear.

If in Fig. 3 the variables are taken to be (σ^β), ε, and the ' previous time ' τ, and sudden application of unit (σ^β) produces creep such that

$$\psi\,\epsilon = \tau^\kappa,$$

it is found, after an integration by parts for stress zero initially, that

$$\psi\,\epsilon = \kappa \int_0^t (t-\tau)^{\kappa-1}\,\sigma^\beta\,d\tau \quad \ldots\ldots\ldots\ldots (8)$$

in which t, the current time, is any particular value of the variable τ. When the variation of stress with time τ is specified, the right-hand member may be directly integrated to provide the resulting strain at any particular current time. Different loading histories now give different results.

The results immediately required are that, for the

three forms of loading here considered, equation (8) reduces to

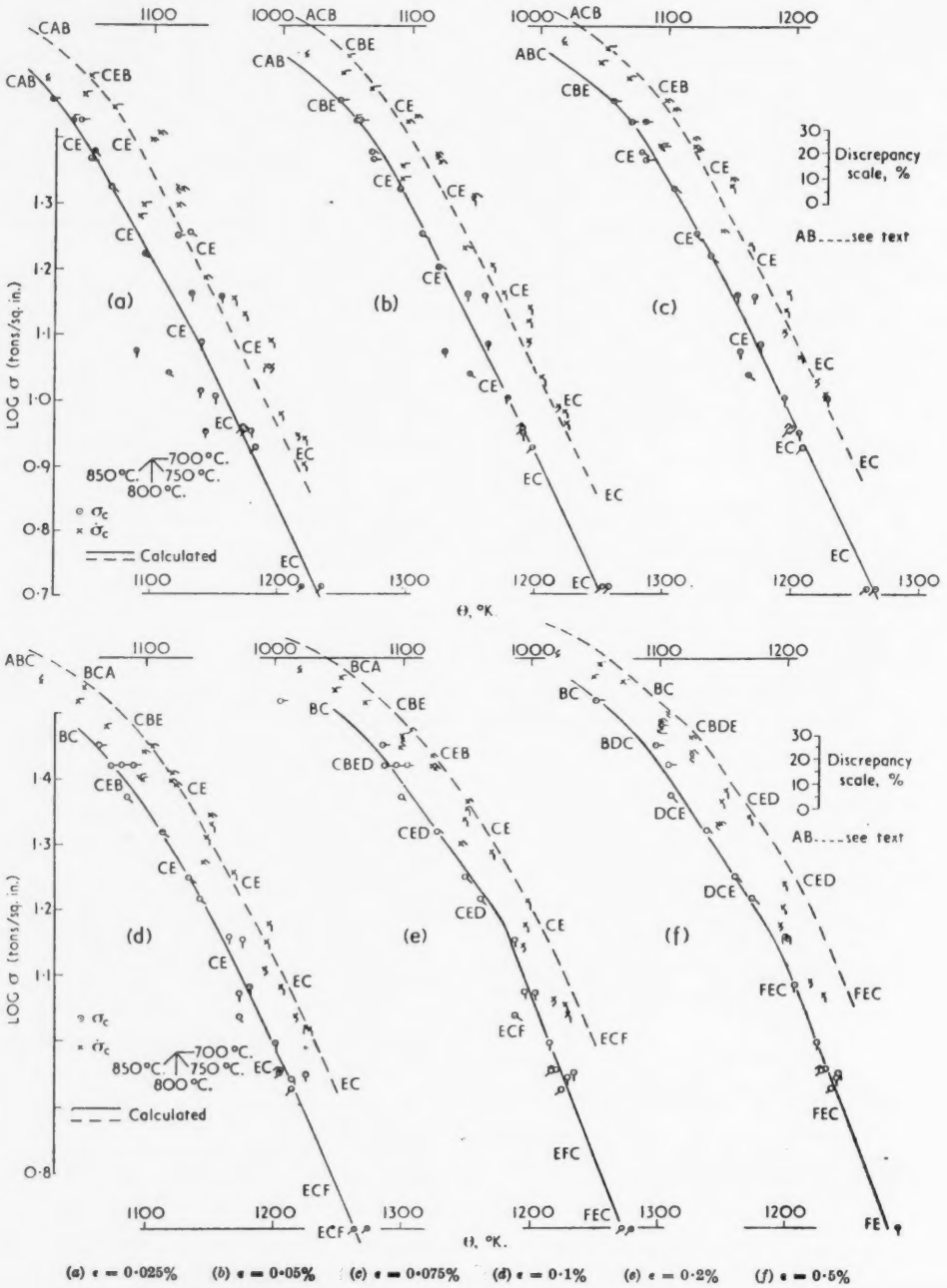
$$\left.\begin{array}{l} \psi\,\epsilon = (K\,\sigma)^{\beta}\,t^{\kappa} \\ \textit{i.e.}, \text{ to the Nutting equation with a factor } K, \text{ in which } K \text{ has the values} \\ \sigma_\epsilon \text{ test: Unity,} \\ \dot{\sigma}_\theta \text{ test: } \kappa\,B\,(\kappa,\ 1+\beta), \quad \text{for } \kappa > 0,\ \beta < 1, \\ \dot{\epsilon}_\epsilon \text{ test: } \kappa\,B\,(\kappa,\ 2-\kappa) \quad \text{for } 2 > \kappa > 0; \end{array}\right\} \qquad \ldots\ldots(9)$$

B is the beta function which may readily be obtained from mathematical tables.

By means of equation (5), the equation may be written

$$\frac{K\sigma}{\sigma_\theta} = \epsilon^{\mu}\,t_\theta^{\zeta\theta}$$

which shows that, in Fig. 1, the lines corresponding



(a) ϵ = 0·025% (b) ϵ = 0·05% (c) ϵ = 0·075% (d) ϵ = 0·1% (e) ϵ = 0·2% (f) ϵ = 0·5%

Fig. 4—Concurrent modes: comparison on basis of θ (log t_θ = −24·8)

to the different forms of loading should all be parallel and be displaced relative to each other by the factors K which are calculable from equation (9) and Table I. The calculated displacements for the decelerating stage are shown by the scale at the side of Fig. 1, and the theory may thus be seen to give a very fair explanation. Application of the factors to the accelerating stage requires a definite standpoint to be taken upon how the two stages are related, and this matter is considered in the next section.

MODES OF DEFORMATION

The other two problems mentioned in the section on " Preliminary Work " are linked by the fact that a formula of the type of equation (4) applies, apparently to within experimental uncertainty, to decelerating and accelerating creep, for both of which stress is constant, and also to any linear section on a diagram like Fig. 2 for which strain is constant. The onset of tertiary creep is generally believed to be associated with a change in ' mode of deformation.' Moreover, Grant and Bucklin[17] and Servi and Grant[18] have clearly shown that the different linear sections in diagrams like Fig. 2 are likewise associated with different ' modes of deformation.' The assumption is clearly indicated, therefore, that all the modes that are involved obey the same formula, namely, equation (8), although each may have its individual values of the constants.

A little thought will show that decelerating and accelerating stages of creep will occur at constant stresses that lie in either the high- or the low-stress sections of Fig. 2, so that at least four different ' modes of deformation ' will be involved. Several of Larson and Miller's and Grant and Bucklin's diagrams show more than two sections, so that, in general, the number of modes is likely to be more than four. This complexity in a complex alloy is not a matter for surprise, in view of the likelihood that each of the several components of its structure may deform in a number of different ways. The simplifying assumption being made is that all ways are governed by a formula of a single type but with differences in the values of the constants.

Much attention had to be given to the next problem, of how the different modes combine together; more specifically, whether one mode of structural deformation ceases completely when another takes its place, or whether all operate together, simultaneously, changing in relative importance as deformation progresses. In practice, there was extremely little to choose between the two possibilities on the basis of the results, but such evidence as there was pointed to the latter more physically reasonable assumption.

The measured strain has then to be regarded as being made up of the sum of contributions like equation (8), each corresponding to a mode, so that

$$\epsilon = \kappa_1 \int_0^t (t-\tau)^{\kappa_1 - 1}\, \sigma^{\beta_1}\, d\tau + \kappa_2 \int_0^t (t-\tau)^{\kappa_2 - 1}\, \sigma^{\beta_2}\, d\tau + \ldots\ldots\ldots\ldots (10)$$

the subscripts indicating the different values of the constants. In this form, the theory accounts for a wide range of known behaviour, but it is not possible in the present paper to enter into details. A summary of predictions is given in the Appendix.

Table III

CONCURRENT MODES: VALUES OF CONSTANTS

N.G.T.E. Material

Mode	$\zeta \times 10^4$ per °C.	μ	Log σ_s	$\dot{\sigma}_c$ Stress-Factor (Δ log σ)
A: Sub-creep	0·04	0·053	1·868	0·006
B: Low θ	0·63	0·053	3·300	0·076
C: Medium θ, primary	1·260	0·262	5·643	0·097
D: Medium θ, tertiary	1·129	$0{\cdot}039_7$	4·612	0·135
E: High θ, primary	1·87	0·318	6·679	0·142
F: High θ, tertiary	2·157	$0{\cdot}072_5$	7·745	0·210

Log $t_s = -24{\cdot}80$ throughout; σ, σ_s in tons/sq. in., t_s in minutes

If attention is confined to σ_c and $\dot{\sigma}_c$ loading, each term in equation (10) reduces to equation (9), giving

$$\epsilon = \frac{(K_1\sigma)^{\beta_1} t^{\kappa_1}}{\psi_1} + \frac{(K_2\sigma)^{\beta_2} t^{\kappa_2}}{\psi_2} + \ldots\ldots\ldots (11)$$

In Figs. 4*a–f*, the complete set of short-time σ_c and $\dot{\sigma}_c$ results from Figs. 1 and 2 are re-analysed on the basis of equation (11). The more scattered $\dot{\epsilon}_c$ results were not considered to be worthy of further analysis. Inclusion with the results of Fig. 1 of the results shown by blocked-in squares in Fig. 2 was found to alter the value of log t_s from $-17{\cdot}3$, as appropriate to Fig. 1, to $-24{\cdot}8$.

The four modes directly indicated by the association of decelerating and accelerating creep with both high- and low-stress regions in Fig. 2 are designated *C–F* in Figs. 4*a–f*. An attempt was made to include, to order of magnitude only, from data given in the Mond booklet, two lower-temperature modes, designated *A* and *B*. Their contribution in the normal creep range is, however, negligible. The constants found are given in Table III.

In the diagrams, the full lines passing through the creep points are calculated from equation (11) and the constants in Table III, which, but for modes *A* and *B*, were derived from these same points; the broken lines represent the corresponding positions, calculated according to these values of the constants, that the $\dot{\sigma}_c$ points should occupy. The principal modes that operate in different parts of the experimental range (but excluding those in the region concerned that contribute less than 10% of the total strain) are indicated in decreasing order of importance. The relative contributions are somewhat different for the two forms of loading, because the correction factors K, though all unity for creep, have, for $\dot{\sigma}_c$ loading, as shown in Table III, other values which differ for the different terms.

The calculated lines are nowhere quite straight because the strain is nowhere due entirely to a single term. The experimental points, however, are too scattered to distinguish straightness from curvature, except in the regions log $\sigma = 1{\cdot}4$ in the $\dot{\sigma}_c$ tests and between $1{\cdot}1$ and $1{\cdot}2$ at the higher strains in the creep tests; in these regions, sharp transitions like

Table IV

CONCURRENT MODES: COMPARISON OF MEASURED AND CALCULATED STRAINS AT ONSET OF TERTIARY CREEP

N.G.T.E. Material

Log σ θ, °K. ϵ_c% calc.	0·7 1260 0·07	0·8 1240 0·09	0·9 1225 0·11	1·0 1210 0·14	1·1 1190 0·17	1·2 1160 0·16	1·3 1125 0·12	1·4 1070 0·05	1·5 0·01
ϵ_c % meas.	0·08		0·11	0·11	0·10	0·16	0·11		0·15
" "	0·03			0·10	0·14	0·16	0·10	0·13	
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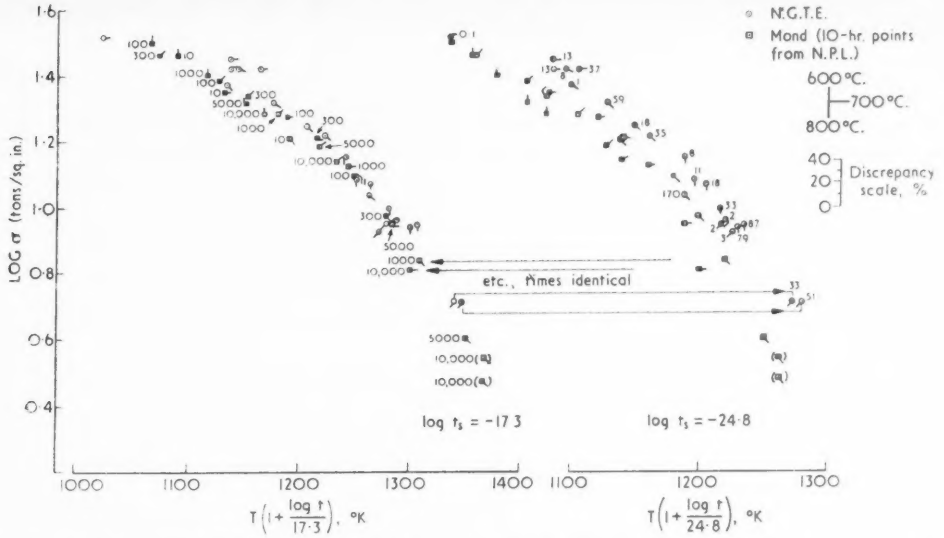

Fig. 5—Comparison of N.G.T.E. and Mond Nickel creep on basis of alternative values of log t_s. Numbers are times in hr., $\epsilon = 0{\cdot}2\%$

accelerating creep as comprising two stages, the Andrade equation, which may be written

$$\epsilon = \epsilon_0 + at^{\frac{1}{3}} + bt \quad \ldots\ldots\ldots\ldots\ldots\ldots (12)$$

in which ε_0 (the 'instantaneous' strain on loading), a, and b are constants in a particular test, merits special attention. For creep at constant stress, equation (11) reduces (*cf.* equation 9) to

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in which D_1, D_2, etc., are constants. If

$$\kappa_1 = 0,\ \kappa_2 = \tfrac{1}{3},\ \kappa_3 = 1,$$

the first three terms are those of Andrade's equation; this is thus a special case of equation (10). According to the present theory, the strain due to the term with zero κ should recover instantaneously, as required by equation (12); that due to the term in $t^{\frac{1}{3}}$ should recover progressively; and that due to the steady-state term should be permanent.

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The Andrade equation does not include accelerating creep, and the addition of another term of equation (13) is therefore necessary. It was found on re-analysis of the data that the fit of the Andrade equation to the decelerating portion of the curve was practically indistinguishable from that of the single term previously used, and the accelerating strain (being that which remained after subtracting the contributions of the previous term or terms) required a substantially unchanged value of κ, namely, about 2·8. Now, if the exponents of two of the terms are in fact the rational quantities $\frac{1}{3}$ and 1, it is very unlikely that the exponent of the accelerating term will be the irrational number 2·8: this number is presumably an approximation to 3. It was found by trial that each of the creep curves of the set could be fitted at least as well by use of terms with the predetermined exponents $\frac{1}{3}$, 1, and 3, as by the previous two terms with freely adjustable exponents, and the new assumption was therefore shown to be justified. The change requires one less adjustable constant to be determined from a curve. Figure 7 is typical of the accelerating creep term according to the original assumption and Fig. 8 according to the new assumption.

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An immediate consequence of this development is that doubt is cast upon the assumptions concerning the influence of temperature, for the exponents κ in equations (6) and (11) have been taken by equation (5) to vary with temperature, whereas those of the Andrade equation are independent of temperature. The experiments provided no information whatever on this matter because the random scatter was much larger than the variation predicted.

When power laws like equation (4) are directly fitted to experimental results, some or all of the constants are not infrequently found to vary with temperature; however, there appears to be no common agreement upon the form of the variation. This is not surprising, for there is abundant experimental evidence (of the nature of graphs with linear sections and abrupt changes of slope) for the general need to represent the relation between σ, ε, t and T by a formula comprising several terms; yet such representation is seldom attempted. In the absence of such an analysis, any apparent temperature variation

Table IV

CONCURRENT MODES: COMPARISON OF MEASURED AND CALCULATED STRAINS AT ONSET OF TERTIARY CREEP

N.G.T.E. Material

Log σ θ, °K. ϵ_c% calc.	0·7 1260 0·07	0·8 1240 0·09	0·9 1225 0·11	1·0 1210 0·14	1·1 1190 0·17	1·2 1160 0·16	1·3 1125 0·12	1·4 1070 0·05	1·5 0·01
ϵ_c % meas.	0·08		0·11	0·11	0·10	0·16	0·11		0·15
" "	0·03			0·10	0·14	0·16	0·10	0·13	
" "				0·10				0·12	
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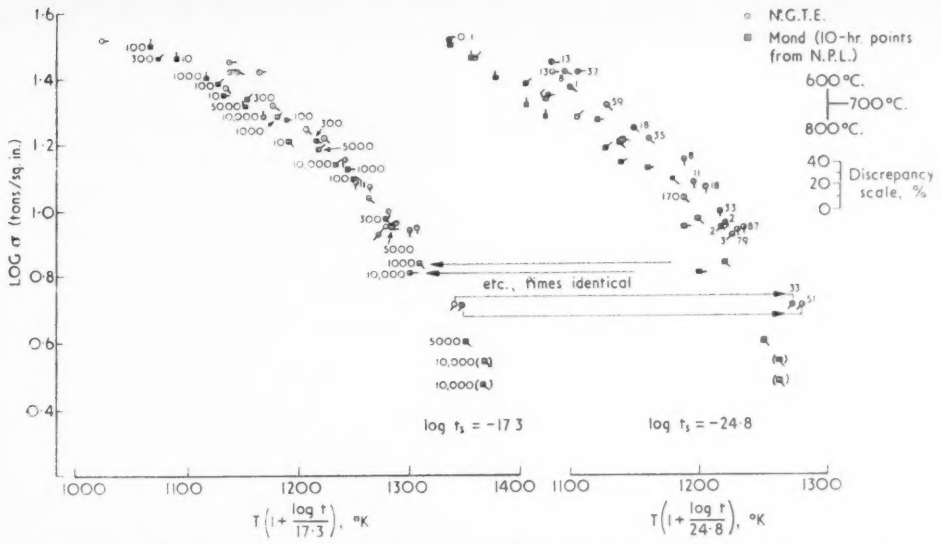

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of an exponent would be without significance. Experience with the present material suggests that extremely few existing data are sufficiently extensive for the separation into terms to be clearly defined. On the other hand, the success of the three-term Andrade equation when applied[22,23] to a wide range of materials at different fractions of their melting points is rather strong support for the constancy of its exponents.

Temperature is commonly thought to enter by way of the Arrhenius factor $e^{-Q/RT}$, the activation energy Q being either a constant or a function of the other variables according to the detailed mechanism assumed. It does not appear always to be realized,

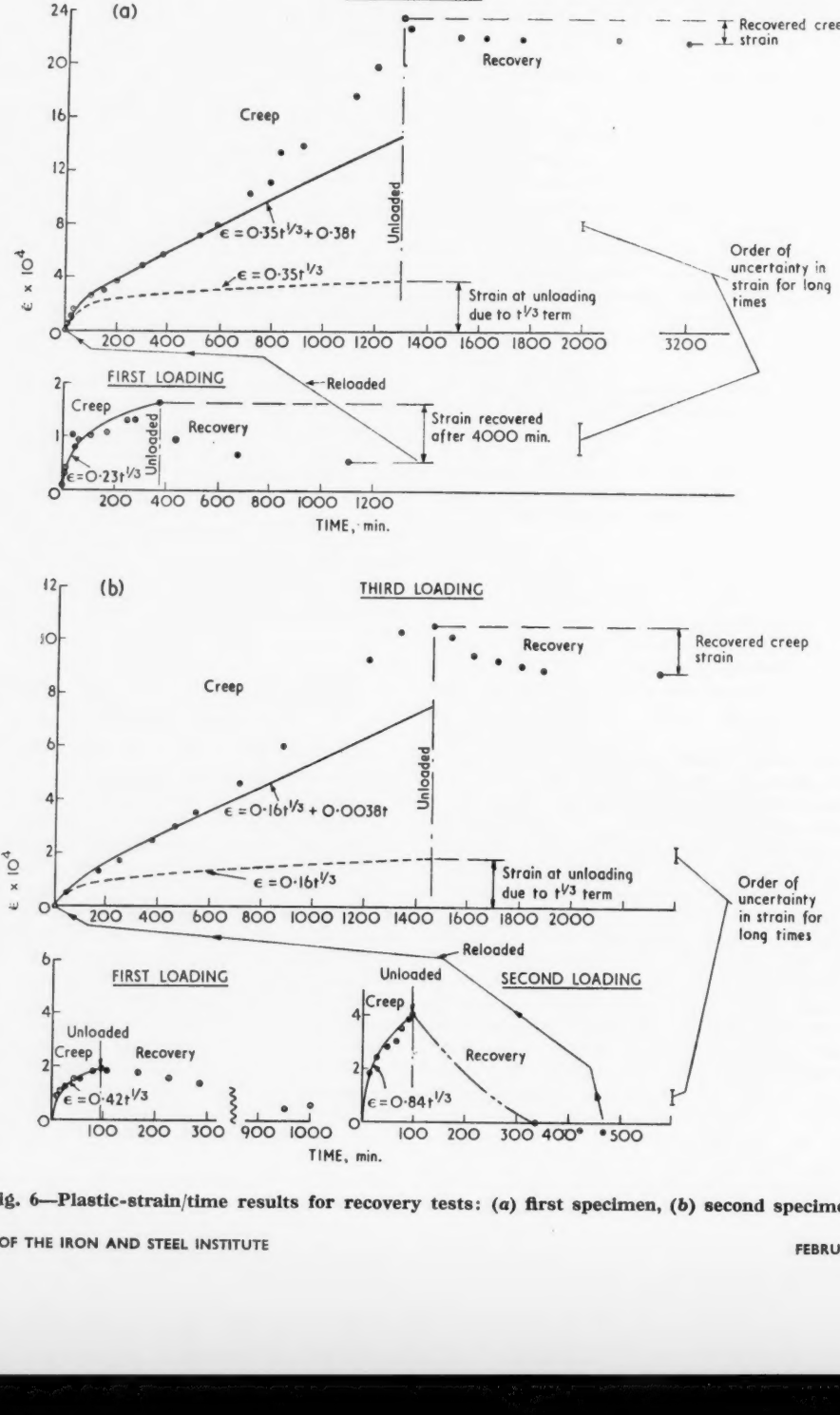


Fig. 6—Plastic-strain/time results for recovery tests: (*a*) first specimen, (*b*) second specimen

however, that the results of Dorn, Goldberg, and Tietz[10] and other similar results completely discredit any assumption of this kind, which results in a theory in which the effect of temperature is *only* represented by a temperature-dependence of the constants. For if, after an experiment has been in progress at a given temperature for some time, a quick change of temperature is made, the only immediate means provided by such theories for calculating subsequent behaviour is the formula as before, but with constants appropriate to the new temperature. The experiment should accordingly proceed thereafter as if the second temperature had prevailed from the beginning. The results mentioned, however, conclusively disprove this prediction.

If the Andrade viewpoint that the exponents are constant is carried over to the more general formulæ (11), (10), and (4), attention is directed to Nutting's suggestion[16] that temperature should appear in his formula symmetrically with respect to the other variables, thus:

$$\psi\, \epsilon = \sigma^{\beta}\, t^{\kappa}\, T^{\nu} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (14)$$

This equation suffers from the objections already mentioned in regard to the history of temperature, just as its forerunner, equation (4), was unsatisfactory in regard to the history of stress; but the remedy may well be similar, namely, to introduce similarly a memory principle. There is no difficulty in doing so in a restricted manner, for, in view of the symmetry of equations (4) and (14) with respect to all the variables, the reasoning of the section on the effect of loading history may be applied as it stands to any other variables with no more than a change of symbols.

An immediate consequence of the introduction of a memory principle is that the material considered both changes with time and is changed by use or by measurements made upon it. The particular case of recovery provides an illustration, for application of load initiates a changing strain which continues, according to the Boltzmann principle, after the load is removed; this strain, together with the subtractive strain initiated by the removal of load, continue their course thereafter and superimpose upon any subsequent strains that would follow from any subsequent applications of load. They thus affect, in a manner that changes with time, the apparent properties of the material. A memory principle in association with temperature will introduce analogous effects in respect of temperature changes.

Machining into a test specimen and putting it under test is only the last of the several changes that a material experiences from the earliest moments of its history, so that under test, or even if it is not tested, the changing traces of its previous history will cause changes of its properties. The present arguments offer the possibility of relating these changes to the previous history. Presumably they will be associated with changes visible under the microscope or detectable by other means, but these questions can be treated entirely separately.

Detailed calculations of the effect of temperature history must await provision of the necessary data, but the following approximate argument appears to

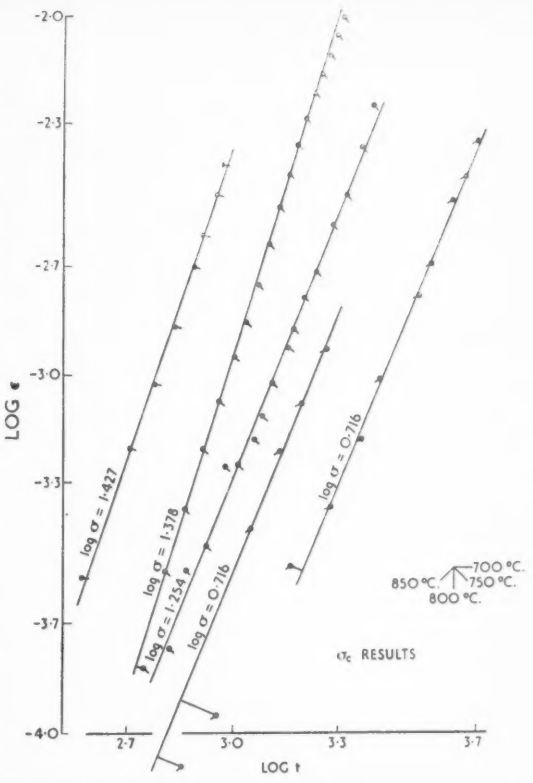


Fig. 7—Accelerating creep when decelerating creep is represented by a single term

meet the requirements of the results. When an equation of the form of equation (8) is used in respect of temperature, the symbols T for the current temperature and, say, λ for the previous temperature replace t and τ; in place of σ^{β}, which, it will be noted, was to be expressed in terms of time τ by consideration of the history of loading, there will occur some functions of stress, strain, and time, expressed in terms of temperature λ by similar consideration of the history of stress, strain, and temperature. This history cannot readily at present be specified in detail but it may be represented, in this position in the formula, to some degree of approximation, by a suitable mean value regarded as constant. If now the material is imagined to be 'born' at some period of high temperature (say, T') during manufacture, the changes will have to be summed over the range from T' to the testing temperature T. By comparison with equation (8), this will give, for the form in which temperature is to occur in the formula,

$$\text{const.} \times \int_{T}^{T} (T - \lambda)^{\nu-1}\, d\lambda,$$

$$i.e., \text{ const.} \times (T - T')^{\nu}.$$

Thus, $(T' - T)^{\nu}$ replaces T in equation (14) and, for creep, a sum of terms of the form

$$\sigma^{\beta}\, t^{\kappa}\, (T - T')^{\nu}. \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (15)$$

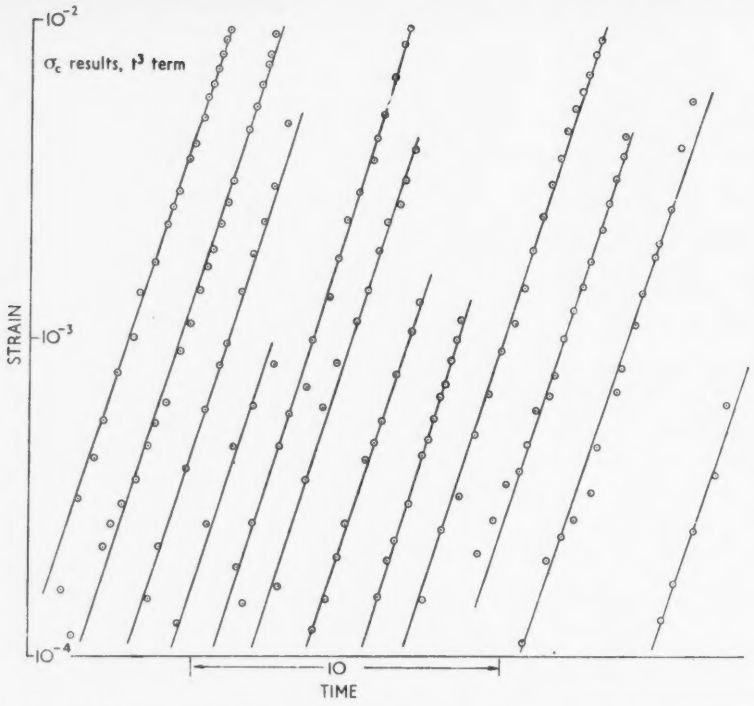


Fig. 8—Cubic accelerating creep when decelerating creep is represented by Andrade terms. Lines are of slope 3

should be appropriate, so that time and temperature may be combined in suitable circumstances into a ' time-modified-temperature ' of the form

$$t^{\kappa}\,(T - T')^{\nu},$$
$$\text{or } [t(T - T')^{-A}]^{\nu} \text{ if } A \equiv -\kappa/\nu.$$

With $T' > T$, a question of sign arises which is to be decided in principle by the magnitude of the appropriate exponent and use of the Argand diagram. The present results, however, require a positive value, and the time-temperature variable ϕ is accordingly taken to be

$$t(T' - T)^{-A} \equiv \phi \qquad \ldots\ldots\ldots\ldots\ldots(16)$$

The total creep strain is then represented (*cf.* equations (16), (15), and (11)) by a sum of a number of terms, one for each mode, the first being

$$C_1\,\sigma^{\beta_1}\,\phi_1^{\kappa_1}.$$

For the present materials, experiment gave no clear indication of a difference between the values of A and T' in the different terms, and economy of hypothesis therefore demanded the use of common values. The quantities ϕ_1, ϕ_2, etc., in the different terms may then also be taken to be common. Of the four modes or terms that were identified, two had the same exponent $\kappa = 3$, so that the equation finally used was, in view of the establishment of the κ values $\frac{1}{3}$, 1, 3,

$$\epsilon = C_1\,\sigma^{\beta_1}\,\phi^{\frac{1}{3}} + C_2\,\sigma^{\beta_2}\,\phi + (C_3\,\sigma^{\beta_3} + C_4\,\sigma^{\beta_4})\,\phi \ldots\ldots(17)$$

FINAL PRESENTATION OF RESULTS

At this stage, the Mond results for tests extending to 22,000 hr. on an early heat Z88 of Nimonic 80A became available for analysis. Owing to the limited range of temperature covered individually by any single set of data, reliable values of A and T' could not be obtained from any one set alone. In view of favourable indications, all data with the exception of that tabulated by Mond Nickel, which was not in a form suitable for direct analysis, was treated for this purpose as a single set. The values $A = 22{\cdot}5$ and $T' = 1175°$ C. were in this way obtained. The assumption that these constants were not only the same for the different modes of any one of the materials, but were also the same for all the materials was justified by the results. Table V gives the values found for all the constants.

Figure 9 for the ϕ^3 term of equation (17), showing $\log \sigma$ *vs.* $\log \phi$ for times at which the term has the constant magnitude 0·1%, is typical of the remaining terms and strains. The line is that calculated for the short-time points from equation (17) and the constants of Table V. The ϕ^3 term, however, is now the only one that shows the θ transition in the neighbourhood of $\log \sigma = 1{\cdot}1$; the transition is seen to occur at the same stress in both short- and long-time results. The slopes of the lines through the individual sets in Fig. 9 closely agree, and similar agreement is found for the other terms, so that since these slopes are $-\kappa/\beta$ and the fixed κ are common to all, the constants β are shown to be common.

In diagrams like Fig. 9 the time involved in ϕ is that for which the *term* considered makes a given contribution to the total strain. Since all data are co-ordinated by use of common values of A and T', it is possible to plot data in the form $\log \sigma$ *vs.* $\log \phi$ in which the time involved in ϕ is that for which the *total strain* ε has a given value. The condition is not necessarily equivalent to that in Fig. 9; however, no essential difference could be established for the present material, and Figs. 10*a–d* give all the data in this form, including now that from the Mond Nickel booklet, for the three constant total strains 0·1% 0·2%, and 0·5%, as in the booklet. The lines are again those for the N.G.T.E. data calculated from equation (11) and Table V; they are representative of those corresponding to the other sets of points.

At stresses below 8 tons/sq. in., the points derived

from the long-time data do not fit well; but the data, especially at small strains, show inconsistencies which, while probably not affecting the validity of the ordinary stress/log-time or log-stress/log-time plots for a given temperature, give rise to difficulty when the variation with temperature is considered.

When the sets of data are considered individually, the new variable ϕ has little advantage over the previously used T_m or θ, but when all the data are considered, the advantage is definite. Common values of the constants in the time-temperature variable are now equally suitable for all, and no discrepancy now exists between the slopes of the lines for the different materials. From the present viewpoint, since values of A, T', κ, and β are common, the series of materials that have resulted from a continuous improvement of the alloy from an early stage of development differ from each other only in a steadily changing trend of the constants C, the later materials being the stronger. The data for the different materials, which are thus satisfactorily co-ordinated, cover a time range of 10^4 to 1, the mean ranges being in the ratio of about 10^3 to 1.

The effect of replacing T_m or θ by the new variable ϕ being hardly appreciable for any one set of data alone, the relationships previously established between the σ_e, $\dot{\sigma}_e$, and $\dot{\varepsilon}_e$ results on the log σ/θ plot remain substantially unaffected by the change.

SUMMARY

It was found that individual creep curves could be fitted by a three-term formula with time exponents $\frac{1}{3}$, 1, and 3 to within experimental error; moreover, the absence of systematic effects when points for different temperatures and durations are compared in diagrams such as Figs. 10*a–d* indicates that the concept embodied in ϕ is as good as the data. The scatter of the points in these diagrams is therefore mainly due to variability of the material, and it is accordingly seen that the theory co-ordinates all the creep data to within their scatter. It also co-ordinates similarly the tension-test and creep-recovery data, except that there is some evidence that the correction factors K are too high for terms with $\kappa > 1$ and that the rates of recovery are too rapid. These discrepancies may indicate the need for a more general theory, the nature of which is referred to by Graham.[12]

The theoretical principles used are able in principle to represent the properties of a material as being dependent upon its history before testing, and also as changing with time whether under test or not, but study of these effects in detail is a separate problem. Some indications of their influence may appear in the

Table V

EXPERIMENTAL VALUES OF CONSTANTS IN EQUATION (17)

Material (*a*): N.P.L. long-time tests
Material (*b*): N.G.T.E. short-time tests
Material (*c*): Mond Z88 long-time tests

Constant	$t^{\frac{1}{3}}$ Term	t^1 Term	t^3 Term (i)	t^3 Term (ii)
κ	$\frac{1}{3}$	1	3	3
β	2·42	5·85	22·1	13·9
Log C: material (*a*)	16·54	50·77	146·82	...
,, ,, material (*b*)	16·14	50·53	145·03	154·74
,, ,, material (*c*)	16·19	49·63	143·88	154·14
A	22·5			
T', °C.	1175			

Stress in tons/sq. in., time in minutes

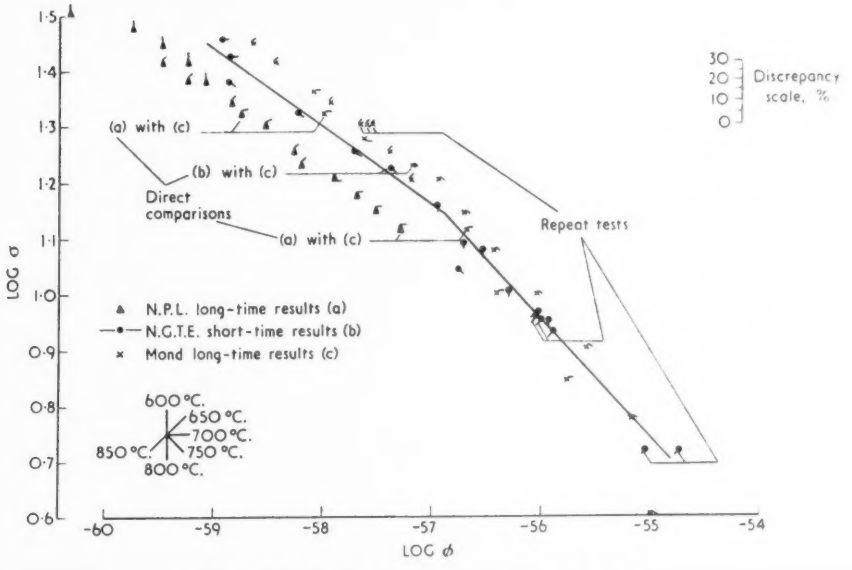
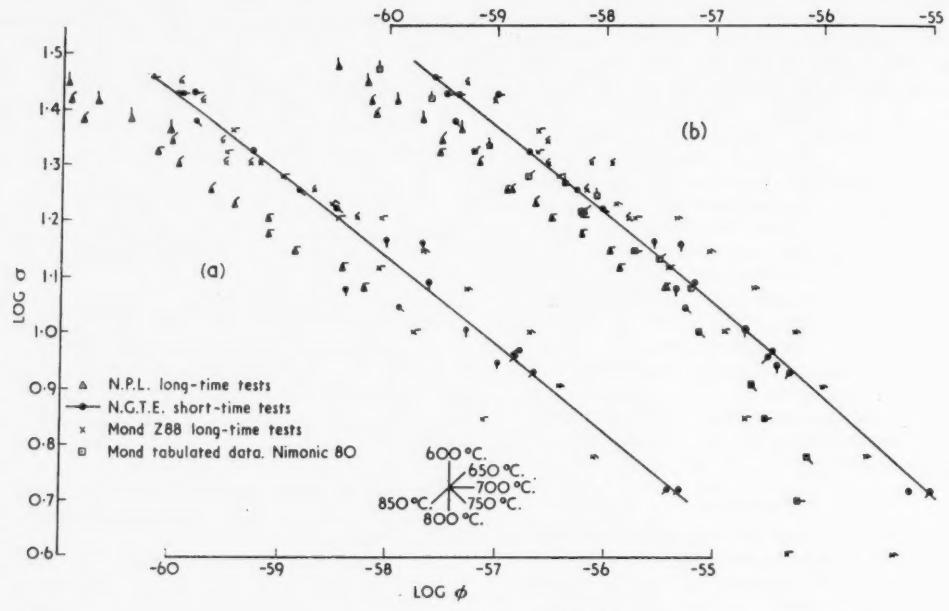
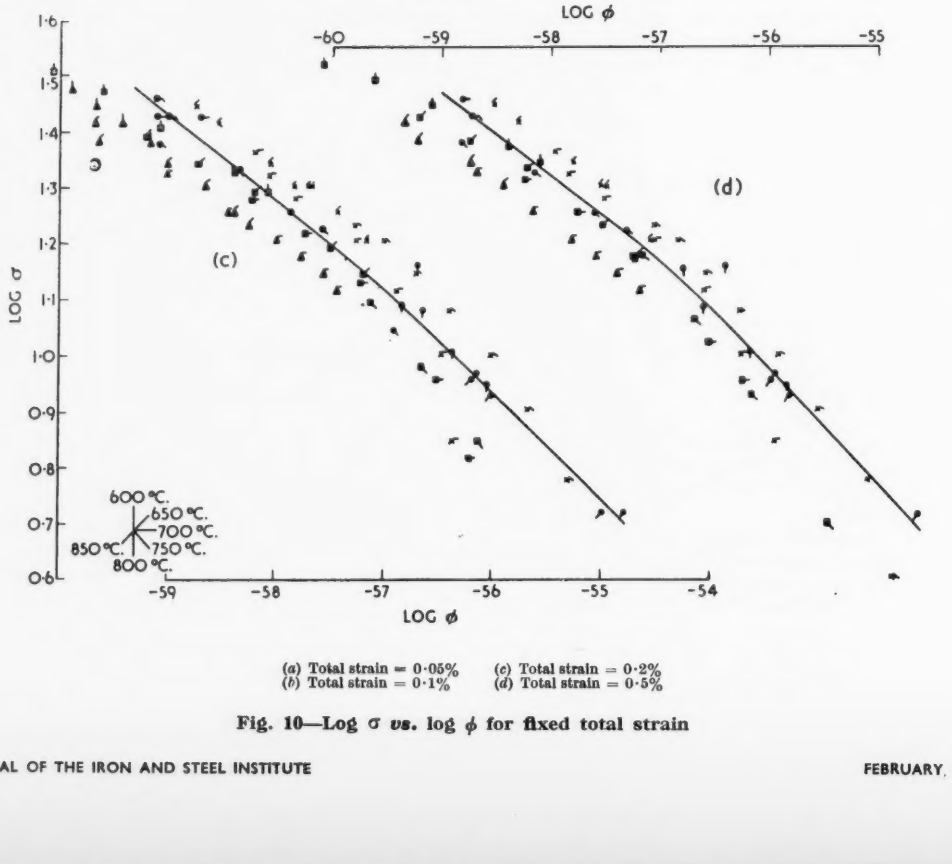


Fig. 9—Log σ *vs.* log ϕ for t^3 term; $\epsilon = 0{\cdot}1\%$

present results, however. In the derivation of the time/temperature relationship ϕ, the material was considered to be 'born' at some temperature T' during manufacture, and there may be some significance in the fact that the value 1175° C. found for T' is about the rolling temperature for the material.

A significant result concerning the change of material with time is that the change of slope of the log-stress/log-ϕ curves has been shown to occur, when due allowance is made for the differences of material, in the same position in both long- and short-time tests. This is conclusive proof that the precipitation





(a) Total strain = 0·05%
(b) Total strain = 0·1%
(c) Total strain = 0·2%
(d) Total strain = 0·5%

Fig. 10—Log σ *vs.* log ϕ for fixed total strain

or other effect which is responsible for the change is subject to the same time/temperature variable as the mechanical properties. The existence of such changes can therefore no longer be held to show the impossibility of predicting long-time behaviour from short-time tests. The present results, in fact, give every indication that such prediction is possible.

The work has shown that, although each of the earlier theoretical possibilities was acceptable for restricted sets of data, deficiencies become apparent when results for different loading histories, time ranges, and materials are compared. The need for a three-term formula for the creep curve with, for the exponents of time, Andrade's values $\frac{1}{3}$ and 1, together with 3 for the remaining tertiary-stage term, rather than a two-term formula with more freely chosen exponents, has in this way been demonstrated. Other experimental evidence exists which suggests that this result applies rather widely, but not universally.

The parameters T_m and θ defined in equations (1) and (2) have by the same means been shown to be approximations which are not good enough for the prediction of long-time data from short-time tests; the new parameter ϕ defined in equation (16) appears to be more promising.

The co-ordination of the various data has involved the freedom of many constants, but these may well be no more in number than is warranted by the complexity of the experimental material. There are also compensatory factors. All the deformation processes that are relevant appear to obey the same formula; the use of two free constants in the time-temperature parameter, equation (16), in place of the one in equations (1) or (2) has led to all the constants except C becoming the same for the four materials. The regular sequence of time exponents also suggests that other constants besides κ may prove to be simply related. The work therefore favours the view that there may be a much greater uniformity between the behaviours of different materials than is generally supposed.

Acknowledgments

The authors wish to thank Mr. H. W. G. Hignett of the Mond Nickel Co., Ltd., and Mr. H. J. Tapsell of the National Physical Laboratory for kindly supplying long-time data.

Acknowledgment is made to the Chief Scientist, Ministry of Supply, for permission to publish this paper. Crown Copyright reserved. Reproduced by permission of the Controller, H.M. Stationery Office.

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APPENDIX

General Predictions of the Theory

A single term of equation (10) predicts the following frequently-observed laws and effects:

The Nutting law (equation (4)) for σ_c, $\dot{\sigma}_c$, and $\dot{\varepsilon}_c$, loading at constant temperature.

This includes, for $\beta = 1$ and $\kappa = 0$, Hooke's Law, and for $\beta = 1$, $\kappa = 1$, the law of viscosity for liquids and gases; intermediate values of β and κ represent intermediate behaviour like that of many real materials.

The following laws of creep:

Creep curves linear on log/log plotting

Curves for different stresses at constant temperature geometrically similar

Log (creep rate) proportional to log (stress) for either t constant or ϵ constant

Linear graphs of log σ *vs.* either T or log t for ϵ constant.

Tensile-test curves of stress *vs.* strain linear on log/log plotting.

With suitable dependence of β and κ upon temperature, either the time/temperature equivalence (equation (2)) or the MacGregor–Fisher form (equation (1)).

Alternatively, with β and κ independent of temperature and with temperature assumed to enter *via* an integral analogous to equation (8), the preferred form equation (16).

Effects of changes of load, straining rate, or temperature, in creep and tensile tests.

Features of stress-relaxation at constant length.

Permanent creep for $\kappa = 1$.

Creep recovery for κ smaller than unity according to the following known laws:

Strain during recovery proportional to maximum forward strain

Time scale during recovery proportional to time of application of load

Recovery curve approximately but not exactly similar to curve of forward creep

Recovery curve approximately but not exactly of power-law form

Increase of temperature during recovery increases rate of recovery.

Hysteresis loops in cyclic loading, constant loops (steady-state cycling). A 'Bauschinger effect.'

The more complete expression (10) with its further terms is potentially able to explain departures from the laws mentioned, and in addition the following:

For terms with suitable exponents, an initial perfectly elastic extension on first application of load (instantaneous recovery if load is removed) followed by 'anelastic' or nearly elastic deformation, recoverable micro-creep (Chalmers), primary, secondary, and tertiary creep, as observed; 'elastic' unloading from the plastic region. Deformation nearly time-independent at low temperatures, and progressively more time-dependent and viscous as either temperature is raised or loading time is prolonged. Abrupt changes of slope as observed when creep data are plotted as log σ *vs.* T, or log t, or θ for constant strain; log σ/log ε curves of segmental form. Reversed-loading hysteresis loops of correct form.

The Andrade equation

$$\epsilon = \epsilon_0 + \beta t^{\frac{1}{3}} + \kappa t,$$

together with his following general results:

$t^{\frac{1}{3}}$ term represents recoverable creep

t term represents permanent creep

ϵ_0 term includes, but may not solely consist of, the elastic strain.

Creep curves of different metals similar at temperatures that are similar fractions of their melting points (T' in equation (16) taken to be T_{mp}).

The above statements are primarily made with reference to metals, but many are known to apply also to non-metals. It is to be clearly understood that no claim is made that all the quantitative laws mentioned are exact statements in themselves of the behaviour of materials, nor that perfect quantitative agreement would be found (if time were available to make the comparisons) where qualitative agreement is implied. The claim is rather that the right kind of predictions are made over this extensive range which includes the main features of tensile deformation.

Effect of Temperature and Pouring Speed on Ingot Structure

By the Ingots Committee of B.I.S.R.A.

SYNOPSIS

A 5-ton high-frequency melt of low-carbon 3% Ni–1% Cr steel was tapped in two halves, with a temperature difference of 40° C. between them; each half was made to produce two 1-ton ingots, which were cast at different rates.

Examination of the ingot surfaces showed the presence of transverse cracks in the ingot cast more rapidly at the higher temperature, whilst that cast more slowly at the lower temperature exhibited a slaggy surface. Longitudinal sections of the ingots showed only a small difference in the structures, the length of the columnar crystals in the ingots cast hotter being, as expected, somewhat longer. The structure revealed by an ammonium persulphate etch, however, was found to be of especial interest; it cannot be accepted as representing the as-cast structure, and a theory is put forward suggesting that it is the result of the peritectic transformation in the particular steel under study. Further work in this connection is being carried out. 1006

DURING consideration of its future programme at a meeting of the Ingots Committee about three years ago, it was generally agreed that the lack of accurate means of measuring liquid steel temperatures had precluded serious study of temperature effects from the numerous researches into structure and segregation carried out on sectioned ingots in the early days of the Ingots Committee (then the Heterogeneity of Steel Ingots Committee). Owing to the advance in recent years of liquid steel pyrometry, the Committee therefore formed a small group* to plan an investigation on structure and segregation in which temperature was the main variable.

METHOD OF INVESTIGATION

The method of investigation finally put forward by this group and subsequently embarked upon was briefly as follows, its attractive feature being that, in addition to using steel at two different temperatures, it practically eliminated the effects of variations in steel composition.

A heat of an ordinary grade of low-alloy steel was to be melted in a basic-lined 5-ton high-frequency furnace. One half of it was to be tapped and teemed at one temperature and the remaining half was to be heated up to a higher temperature and then tapped and teemed. In regard to uniformity of composition between the two halves of the charge, it was realized

* The members of the group were: Mr. K. C. BARRACLOUGH, Mr. D. W. HAMMOND, Mr. W. H. STRATFORD, Mr. R. LAMB, Dr. H. H. BURTON, and (later) Mr. H. ALLSOP.

Paper SM/B/15/54 of the Ingots Committee of the Steelmaking Division of the British Iron and Steel Research Association, received 18th May, 1954.

that some difficulty might be experienced in obtaining similar and sufficiently high sulphur contents (say, 0·03–0·04%), and so it was agreed that additions of ferrous sulphide should be made during tapping.

To allow a study of the effect of teeming speed to be included in the investigation, it was decided to subdivide the two halves of the charge into two parts by the use of large- and small-diameter nozzles during teeming. Thus there were to be four experimental ingots* of the same composition, representing four conditions of treatment:

Low-temperature steel	slowly	poured	(*A*)
,, ,,	rapidly	,,	(*B*)
High- ,,	slowly	,,	(*C*)
,, ,,	rapidly	,,	(*D*)

MANUFACTURE OF THE FOUR EXPERIMENTAL INGOTS

As nickel–chromium case-hardening steel of En. 36 type shows a tendency to transverse cracking in the ingot, particularly when the casting temperature is slightly higher than normal, this quality was chosen for the experiment, the melting history of the cast used being as follows:

Experimental H.F. Cast 41478

8.20 a.m.

0·10% C tubes	10,800 lb.	Charged with 450 lb. of lime and 80 lb. of scale
Nickel	329 lb.	

11.20 a.m.

Slag removed		Charged with 200 lb. of lime and 50 lb. of spar
Ferro-silicon (45%) to kill	56 lb.	
0·5% C ferro-chromium	95 lb.	
Aluminium	1 lb.	
Iron phosphide (25% P)	13½ lb.	

11.45 a.m.

Pass test: C, 0·05–0·06%; Mn, 0·14%; Ni, 3·13%; Cr, 0·65%.

12.25 p.m.

Slag removed		Charged with 180 lb. of lime and 60 lb. of spar.
6% C ferro-manganese	45 lb.	
0·1% C ferro-manganese	23 lb.	
6% C ferro-chromium	59 lb.	
Ferro-silicon (45%)	10 lb.	
Calcium silicide on slag	10 lb.	

12.48 p.m.

Approx. 50 cwt. tapped

Slag held back, 7 lb. of iron sulphide added, 60 lb. of broken glass on top of metal — **Cast *A***

Furnace temperature Q.I.P.	1614° C.
(Corrected) optical reading at tap	1595–1600° C.
Ladle temperature Q.I.P.	1575° C.
(Corrected) optical reading, casting	1538–1535° C.
Ingot *A* cast through 1-in. nozzle; time to joint	70 sec.
Ingot *B* cast through 1¾-in. nozzle; time to joint	30 sec.
Spare metal run into 4-cwt. moulds; two made: one remelted for second cast, one reserved	

1.15 p.m.

Remainder tapped

Slag held back, 7 lb. of iron sulphide added, 60 lb. of broken glass on top of metal — **Cast *B***

Furnace temperature Q.I.P.	1650° C.
(Corrected) optical reading at tap	1640–1645° C.
Ladle temperature Q.I.P.	1613° C.
(Corrected) optical reading, casting	1578–1572° C.
Ingot *C* cast through 1-in. nozzle; time to joint	69 sec.
Ingot *D* cast through 1¾-in. nozzle; time to joint	30 ,,
Spare metal run into 4-cwt. moulds, one reserved.	

Analyses

	Cast *A*	Cast *B*
C, %	0·11	0·12
Si, %	0·25	0·23
Mn, %	0·59	0·59
S, %	0·041	0·041
P, %	0·041	0·042
Cr, %	1·03	1·06
Ni, %	3·28	3·28
Mo, %	0·01	0·01
Cu, %	0·10	0·10
Hydrogen, ml. per 100 g.	4·0	4·2
Nitrogen, %	0·009	0·012
Oxygen (Herty), %	0·012	0·009

Material Available

(1) Ingots *A*, *B*, *C*, and *D*: These were cast in standard solid-bottomed moulds, the ingots weighing approx. 19½ cwt. each, with dimensions as follows:

Top of chill	13¼ in. sq.
Bottom of chill	11½ in. sq.
Length of chill	50 in. approx.
Length of head	10–12 in.

(2) Spare pieces (without head) approx. 8 in. sq. (weight about 4 cwt.) from each cast.

EXAMINATION OF THE INGOTS

Inspection of the ingots after stripping showed distinct differences in the surface conditions of the four ingots. Ingot *A* had a slaggy surface free of cracks, ingots *B* and *C* had surfaces representative of normal shop practice, and ingot *D* had numerous cracks in its surface. The appearance of the ingots can be seen in Fig. 1.

After being photographed, the ingots were softened by tempering at 650° C. to facilitate cutting. Then each ingot, including the head, was sectioned longitudinally in such a way that the finished surface for examination would coincide with the central plane. The central longitudinal sections (or half-ingots) were sulphur-printed, macro-etched, and photographed. Analyses taken at the 'standard Ingots Committee positions' are given in Table I.

Data on the mechanical properties of the 'cold' and 'hot' halves of the high-frequency cast, which might be expected to resemble those of ingots *A* and *D*, were obtained by rolling the representative spare pieces into 2 in. square billets, blank carburizing for 6 hr. at 927° C., and oil hardening from 880° C., followed by mechanical testing. The results, which showed no marked differences for the properties of the ingots, were:

	41478*A* (cold)	41478*D* (hot)
Limit of proportionality, tons/sq. in.	18·8	15·2
0·05% proof stress, tons/sq. in.	37·2	33·6
0·10% proof stress, tons/sq. in.	44·0	41·2
0·50% proof stress, tons/sq. in.	60·0	58·8
Maximum stress, tons/sq. in.	74·2	73·0
Elongation, %	16·5	18·0
Reduction in area, %	53·0	54·5
Izod value, ft. lb.	15, 16, 16	17, 19, 16

* The making of the four ingots and the metallurgical examination of them was undertaken by Thos. Firth and John Brown, Ltd., and Brown Bayley Steels Ltd., respectively.

Examination of the longitudinal sections showed that all the ingots were sound in the centre, and there was no evidence of secondary shrinkage or other cavities along the ingot axis. Full-section sulphur prints showed no evidence of abnormal sulphur distribution in any of the ingots. Small spot segregates were visible in all four ingots and were irregularly distributed up to a depth of about 1 in. below the ingot surface. No significant differences were noted on the prints between the four ingots.

The prepared section faces were etched with cold 10% aqueous ammonium persulphate, and an unusual crystalline structure was developed on all four ingots. This structure seemed to consist of three columnar zones with only a very small central core of free crystals. From the surface of the ingots inwards, the zones appeared to be, first, a thin shell of fine crystals essentially at right-angles to the ingot skin, followed by a very coarse crystalline zone of considerably greater depth than the outer zone, and, finally, a series of narrow columnar crystals normal to the ingot surface. Photographs of the full ingot sections were taken; that for ingot *A* is shown in Fig. 2, and is typical of all four sections.

The ingots were inspected by the Committee at this stage and the structures were considered to be so unusual as to warrant further investigation. It should be pointed out that ammonium persulphate has previously been used as an etchant for ingot structures and has revealed columnar and free crystal zones normal for these subjects.

The further examination agreed upon was that the spare halves of ingots *A* and *D* (cold metal slowly poured and hot metal rapidly poured, respectively) should be cut to give the following transverse slices, each 1 in. thick:

Two slices from the top, the top slice being located 1 in. below the shoulder of the ingot
Two slices from just above the bottom of the ingot
Two slices from the middle of the ingot.

The lower slice of each pair was to be normalized at 950° C., the top surface of these slices then being prepared for etching. The lower surface of the upper

Table I

ANALYSES AT 'STANDARD INGOTS COMMITTEE' POSITIONS

Position *	C, %	Si, %	Mn, %	S, %	P, %	Cr, %	Ni, %	Mo, %
Ingot *A*								
A (i)	0·11	0·23	0·59	0·042	0·037	1·02	3·29	0·03
(ii)	0·12	0·24	0·60	0·039	0·042	1·03	3·31	0·01
B (i)	0·10	0·23	0·58	0·041	0·034	1·02	3·26	0·02
(ii)	0·11	0·23	0·60	0·037	0·040	1·01	3·28	0·01
C (i)	0·10	0·23	0·57	0·036	0·031	1·02	3·22	0·02
(ii)	0·11	0·25	0·59	0·033	0·036	1·00	3·23	0·01
D (i)	0·10	0·24	0·57	0·032	0·031	1·02	3·19	0·02
(ii)	0·11	0·24	0·58	0·030	0·030	0·99	3·17	0·01
F (i)	0·10	0·24	0·59	0·038	0·036	1·02	3·29	0·02
(ii)	0·11	0·24	0·59	0·039	0·037	1·01	3·26	tr.
G (i)	0·10	0·24	0·59	0·041	0·036	1·03	3·30	0·02
(ii)	0·12	0·25	0·60	0·040	0·039	1·00	3·29	0·01
H (i)	0·14	0·26	0·62	0·073	0·054	1·08	3·52	0·03
(ii)	0·16	0·24	0·66	0·069	0·056	1·06	3·51	tr.
Ingot *D*								
A (i)	0·11	0·23	0·59	0·044	0·039	1·04	3·32	0·02
(ii)	0·13	0·20	0·61	0·044	0·038	1·04	3·37	0·02
B (i)	0·11	0·24	0·57	0·039	0·035	1·06	3·25	0·02
(ii)	0·11	0·22	0·60	0·037	0·038	1·03	3·27	0·02
C (i)	0·10	0·22	0·56	0·033	0·033	1·04	3·21	0·03
(ii)	0·11	0·22	0·59	0·033	0·031	1·03	3·26	0·02
D (i)	0·10	0·23	0·57	0·034	0·032	1·04	3·20	0·02
(ii)	0·11	0·22	0·59	0·033	0·033	1·03	3·23	0·02
F (i)	0·10	0·24	0·57	0·039	0·036	1·05	3·27	0·02
(ii)	0·12	0·20	0·59	0·042	0·037	1·03	3·31	0·02
G (i)	0·11	0·24	0·60	0·044	0·039	1·04	3·32	0·02
(ii)	0·13	0·22	0·61	0·038	0·044	1·04	3·34	0·02
H (i)	0·14	0·25	0·63	0·059	0·052	1·08	3·48	0·02
(ii)	0·16	0·24	0·64	0·061	0·056	1·08	3·53	0·02

* (i) and (ii) refer to figures obtained by the two different laboratories.

A. Low-temperature steel, slowly poured
B. Low-temperature steel, rapidly poured
C. High-temperature steel, slowly poured
D. High-temperature steel, rapidly poured

Fig. 1—The four test ingots

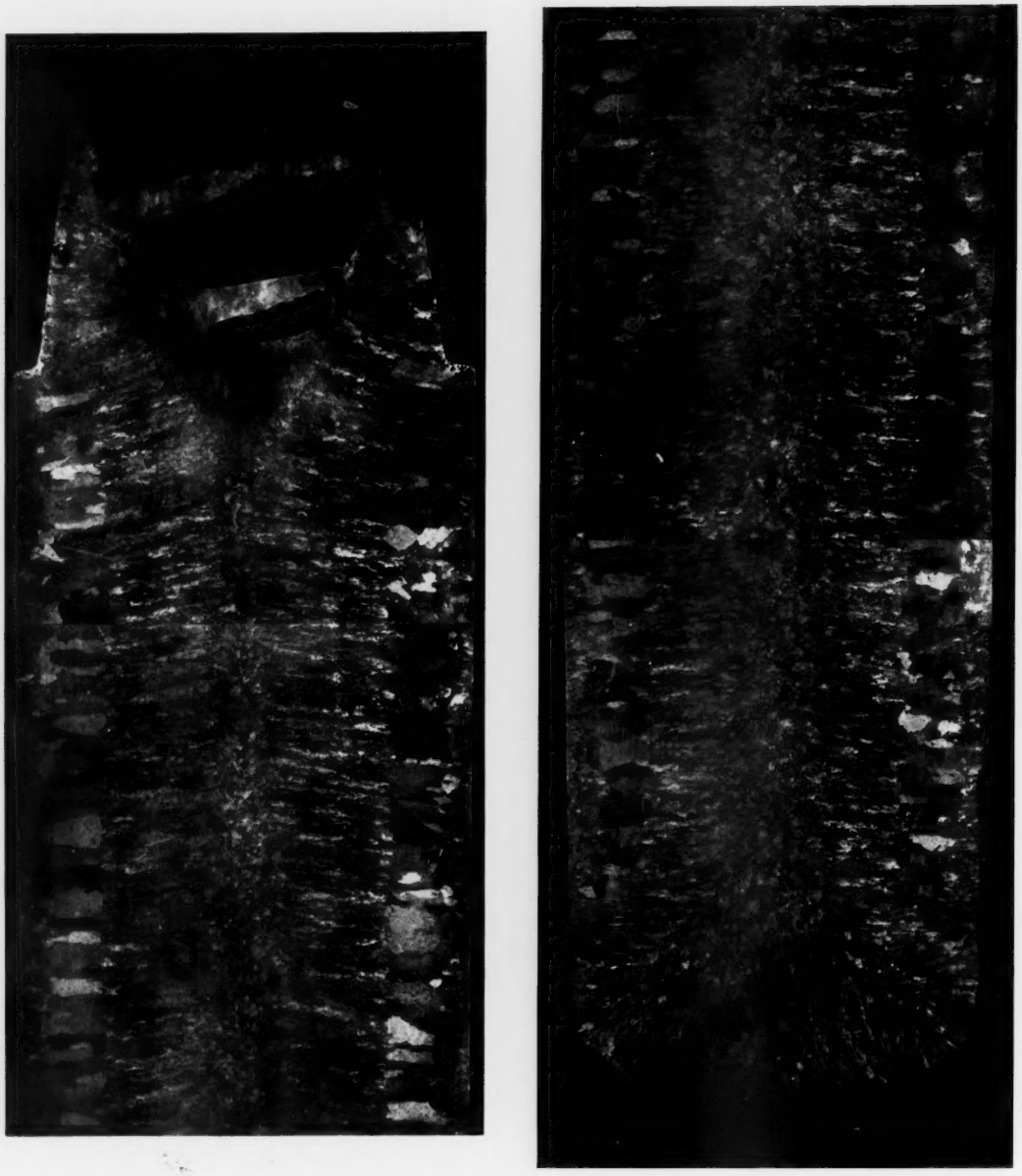
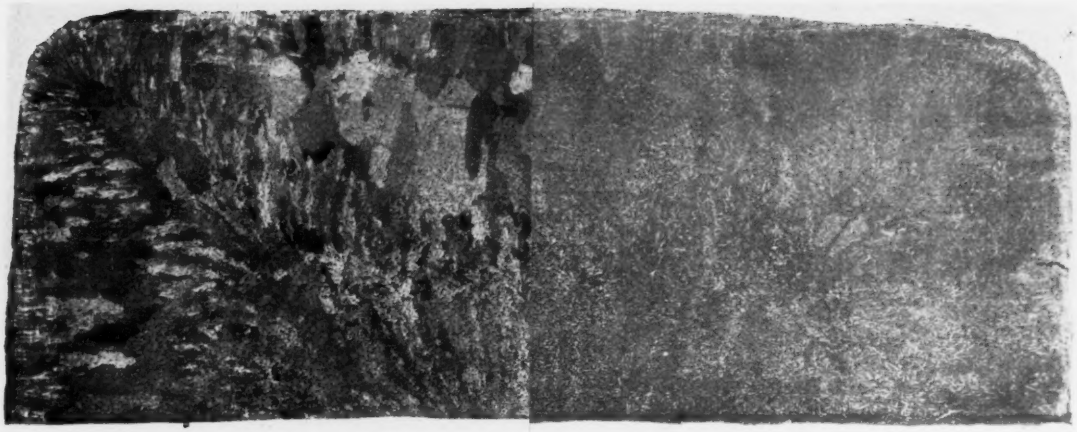
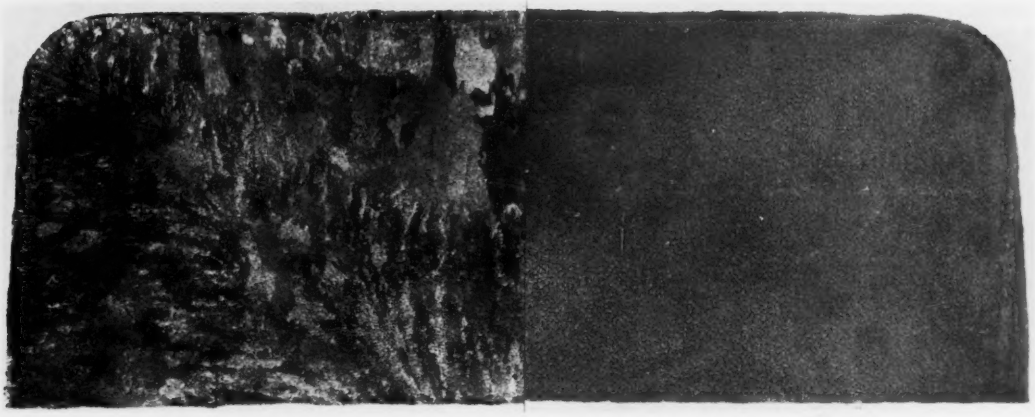


Fig. 2—Ingot *A*: longitudinal section through centre after etching with aqueous ammonium persulphate



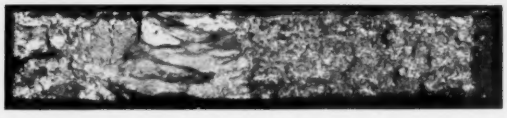
(*a*) Etched with ammonium persulphate

(*b*) Etched with ferric chloride

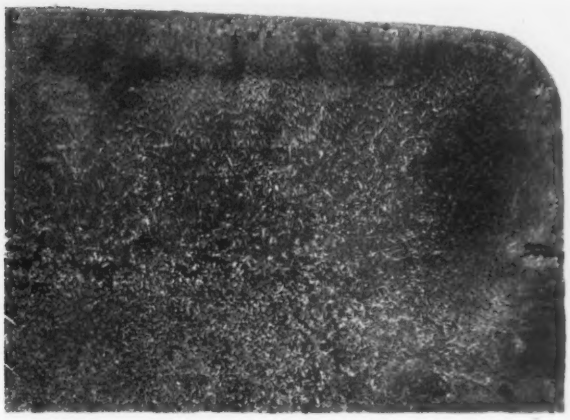


(*c*) Etched with copper ammonium chloride

(*d*) Etched with copper ammonium chloride (contact print)

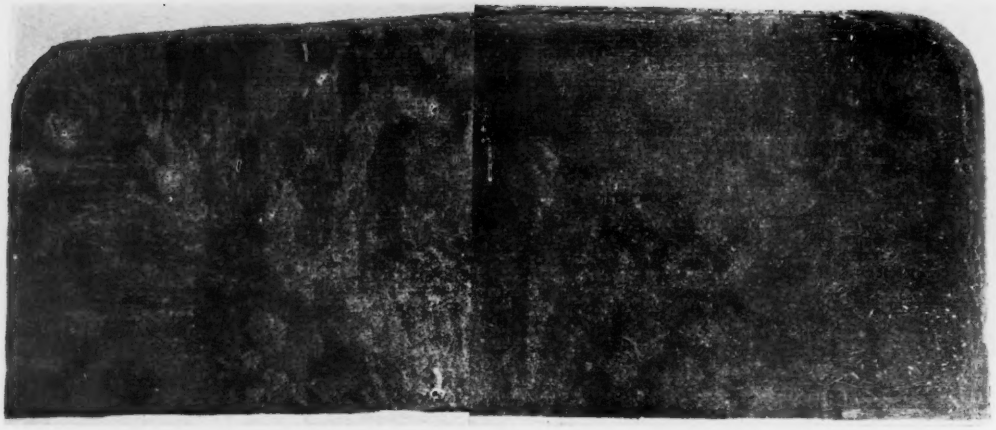


(*e*) Fracture of slice representing edge to centre of ingot



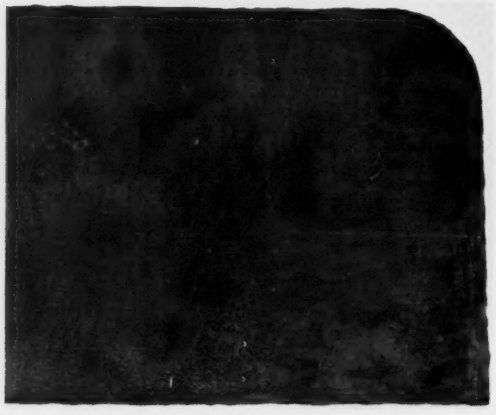
(*f*) Hot etch with aqueous solution of HCl and H_2SO_4

Fig. 3—Structure of ingot *D*, middle, as cast

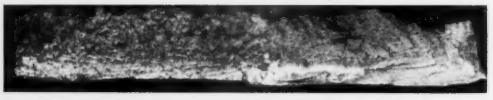


(*a*) Etched with ammonium persulphate

(*b*) Etched with ferric chloride



(*c*) Etched with copper ammonium chloride

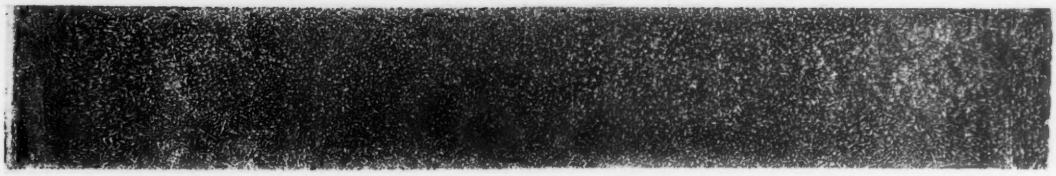


(*d*) Fracture of slice representing edge to centre of ingot

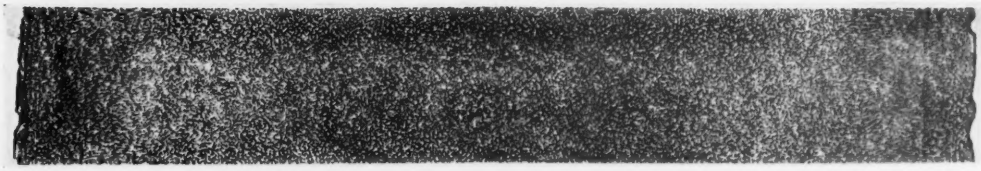
(*e*) Hot etch with aqueous solution of HCl and H_2SO_4

Fig. 4—Structure of ingot *D*, middle, normalized

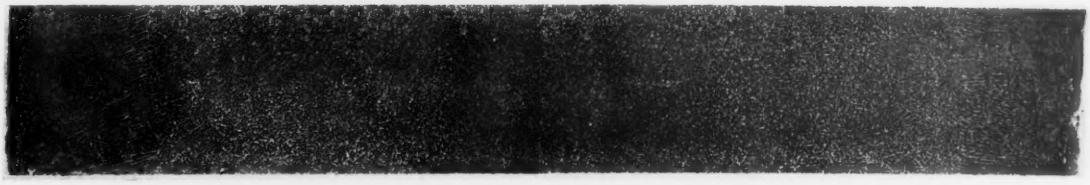
(*a*) Ingot *A* : top



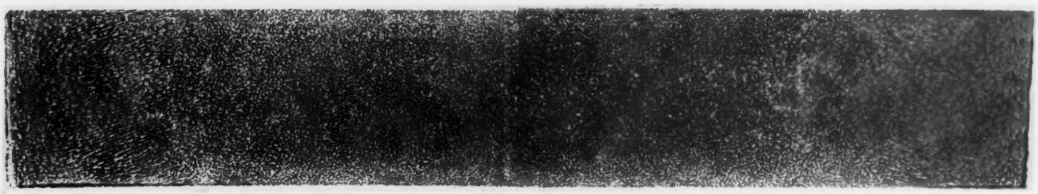
(*b*) Ingot *A* : middle



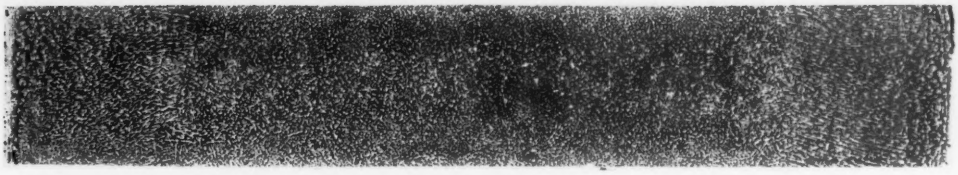
(*c*) Ingot *A* : bottom



(*d*) Ingot *D* : top

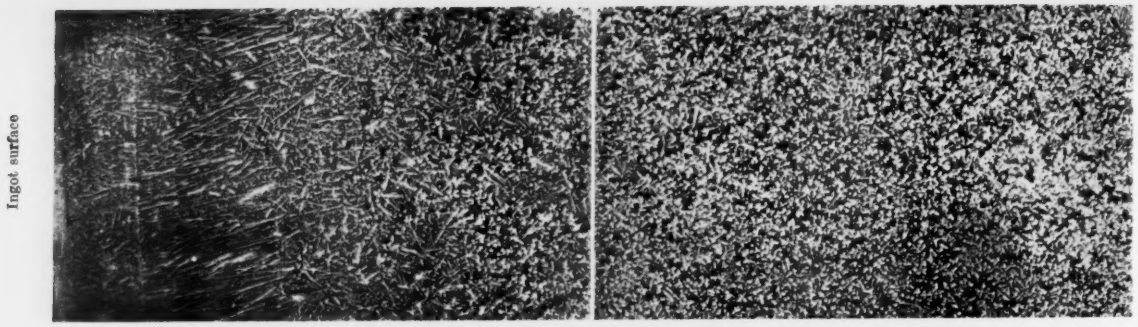


(*e*) Ingot *D* : middle

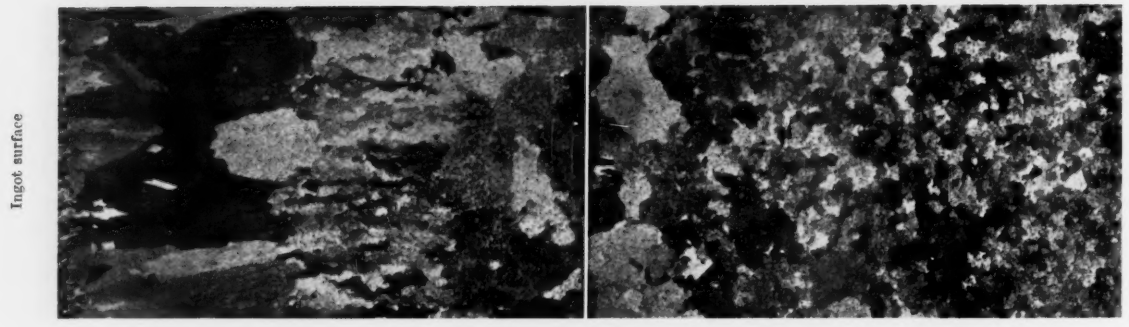


(*f*) Ingot *D* : bottom

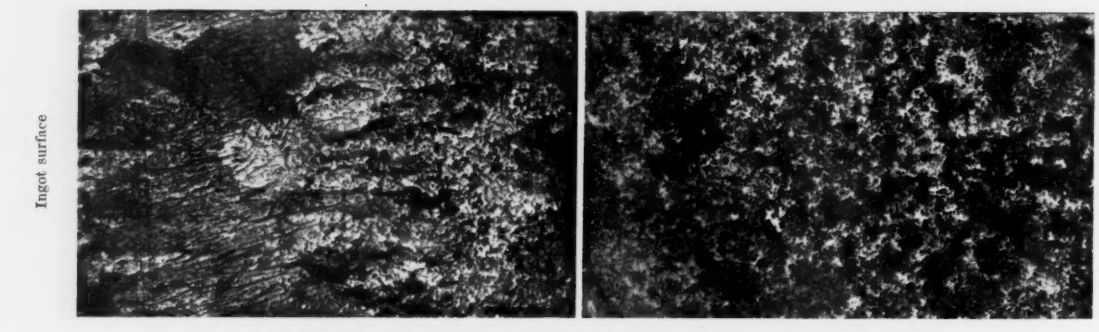
Fig. 5—Contact prints of longitudinal sections of transverse slices after etching, to show dendritic structure



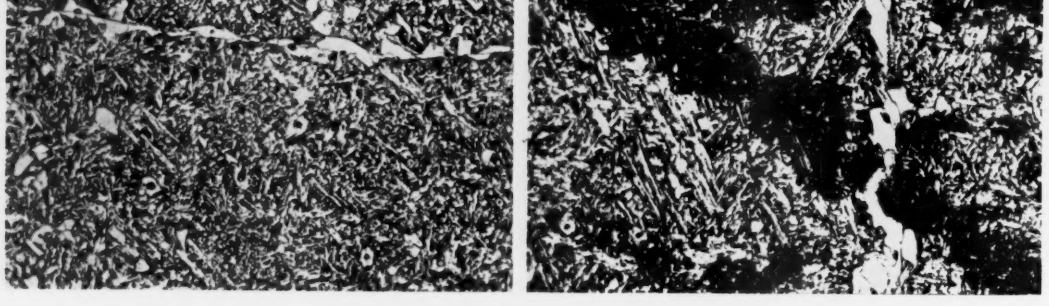
(*a*) Primary structure × 1



(*b*) Secondary structure × 1



(*c*) Primary and secondary structure × 1



(*d*) Tertiary structure with secondary grain boundary × 100 (*e*) Tertiary structure with secondary grain boundary and primary dendrites × 100

Fig. 6—Longitudinal section *AM*, taken from half-way down ingot *A* (low temperature, slowly poured) after normalizing

slice from each pair was also to be prepared for etching, the following etchants to be used:

(i) Ammonium persulphate aq. 10%
(ii) Copper ferric chloride 40 g.–$FeCl_3$, 3 g.–$CuCl_2$, 40 ml.–HCl, 500 ml.–H_2O
(iii) Copper ammonium chloride: aq. 8%; initially neutral, later acidified with 5% HCl.
(iv) A hot etch:

H_2SO_4	1 part by volume	used at 80° C.
HCl	6 ,, ,, ,,	
H_2O	2 ,, ,, ,,	

After etching, each slice was to be fractured through the centre line and the fractures compared.

This procedure was carried out and the structures produced by the various etching media, together with the fractures, were examined and recorded for each of the test positions in ingots *A* and *D*. Independent of the ingot or position in the ingot, each etching reagent produced its characteristic structure, and Figs. 3*a–f*, representing the middle position of ingot *D*, are typical of all six test positions.

A series of photographs representing the same test position as those shown in Fig. 3 (*i.e.*, ingot *D*, middle) but in the normalized condition are shown in Figs. 4*a–e*.

Since in ingots cast in chill moulds the growth of the columnar dendrites is not quite normal to the chilling face, transverse ingot sections are unlikely to give an exact measure of the dendrite length. For this reason, longitudinal sections were etched so as to reveal the dendritic structure and these are shown in Fig. 5, from which it is evident that the columnar growth is consistently greater in ingot *D*, (hot-cast, rapidly poured) than ingot *A* (cast cooler and slowly poured).

Commercial steels solidify over a freezing range, and this generally enables the mode of crystallization to be followed by means of the coring which results from the difference in composition across any one crystal, the coring effect being revealed by suitable etching of a polished section. The form of the crystals may be columnar, equiaxial, or a mixture of the two, the relative proportions being dependent upon the temperature gradient existing during solidification, and also upon the composition of the metal. This dendritic or cored appearance indicates the manner of solidification and is therefore a solidification or primary structure, and in this respect the photographs in Fig. 5 clearly show the effect of casting temperature and speed on columnar growth.

That being so, it is of interest to consider the origin of the crystalline etch developed by the ammonium persulphate solution, as it appears to be independent of the dendritic etch, and especially as, on other occasions with ingots of different composition, it has shown a crystal pattern corresponding with the usual columnar and equiaxial zones normally existing in chill-cast ingots.

Iron–carbon alloys containing about 0·1–0·5% C solidify in part as δ iron, which then reacts with the remaining liquid to transform, at the peritectic temperature, to yield the solid γ form. With higher, carbon contents there is no intermediate δ phase and the liquid steel solidifies directly to the γ condition. On further cooling, both types subsequently transform to the α condition. It follows, then, that with steels of carbon contents within the limits given, which may well be modified by alloy additions, there is an additional phase change, and it is considered that the pattern revealed by the ammonium persulphate etch is a secondary structure formed by the recrystallization of the steel from the δ to the γ form, the boundaries of the grains shown by the secondary structure delineating the austenite crystals. That such a structure and its effect on the fracture should be removed by normalizing is not at variance with this suggestion.

In support of this belief a series of photographs is shown in Fig. 6 which, in addition to illustrating the features shown in Fig. 3, show the final structure due to the transformation of the γ to the α condition. They also show the primary ferrite outlines of the previous austenite grains and, moreover, indicate that these primary ferrite outlines cross the cored pattern of the primary structure.

SUMMARY AND CONCLUSIONS

This work, carried out under the auspices of the Ingots Committee, was intended to investigate the effect of casting temperature on the structure and segregation in steel ingots, primarily because improvements in the means of measuring liquid steel temperatures should permit a better correlation than higherto between the casting temperature and the effects observed. On the cast of nickel–chromium case-hardening steel used, there was a measured difference of 40° C. between the hot-cast ingots and the cool-cast, but the only significant difference in structure was a small difference in the length of the columnar dendrites, those of the hotter-cast ingots being the longer. It was also observed that the hot-cast ingot, cast quickly, exhibited transverse cracks and that the cool ingot, cast slowly, had a slaggy surface. In respect of the purpose of the investigation, therefore, very little information was obtained, and it is considered that any further investigation should be on a larger scale and might to advantage include ingots of higher carbon content.

In the course of the investigation, however, a very interesting effect was produced by the ammonium persulphate etch. The structure revealed by this etchant cannot be accepted as representing the original as-cast structure, and a theory has been put forward suggesting that it is the result of the peritectic transformation in the particular steel under study. Further work on the occurrence of the peritectic change and the structures resulting from it is being carried out.

Acknowledgments

The ingots used in this investigation were provided for the Committee by Thomas Firth and John Brown Ltd., and were sectioned, etched, and examined in the laboratories of Brown Bayley Steels Ltd., by permission of Dr. C. Sykes and Mr. H. Bull, respectively. The investigation was carried out on behalf of the Ingots Committee by Mr. K. C. Barraclough and Mr. H. Allsop of the foregoing firms, respectively. The suggested explanation for the structure shown by the persulphate etch is due to Dr. L. Northcott, as are the photographs shown in Fig. 6.

Deformation of Iron Crystals by Unidirectional Abrasion

By R. P. Agarwala, Ph.D., and H. Wilman, D.Sc., Ph.D., F.Inst.P.

SYNOPSIS

The surface deformation caused on iron crystals by a single 10-in. abrasion stroke upon 0000 emery with light hand pressure was investigated. The electron-diffraction patterns obtained after various degrees of etching showed first, an outermost region of randomly disposed α-iron fragments ; next, a region of random γ iron, followed by an extensive layer of α iron in a wide range of orientations related to that of the main crystal by a rotation about a well defined axis ; and finally, an orientated γ-iron layer in contact with the underlying practically undeformed main α-iron crystal.

On one crystal, which had an approximately (107) surface, the abrasion along directions near $[\bar{1}10]$, $[\bar{2}10]$, and $[70\bar{1}]$ caused rotational disorientations about the [110], [120], and [010] axes, respectively, but abrasion along $[\bar{7}01]$ apparently caused parts of the lattice to rotate about [451], whilst others rotated about $[45\bar{1}]$. On the other crystal, with nearly (110) surface, abrasion along $[\bar{1}1\bar{2}]$, a direction 8° from $[\bar{1}\bar{1}2]$ on the side towards $[00\bar{1}]$, and along $[\bar{1}11]$ caused rotations about $[1\bar{1}\bar{1}]$, $[11\bar{1}]$, and practically $[\bar{3}3\bar{4}]$, respectively. Most of these axes are normal to densely populated lattice planes and could correspond to a lamellar fragmentation involving rotational slip. The [451] axis could, however, correspond to flexural translational slip on $(\bar{2}13)$ along $[1\bar{1}1]$, whilst the $[\bar{3}3\bar{4}]$ seems unlikely to be due to rotational slip, nor does simple flexure explain it if slip is still along ⟨111⟩, as is probable at the temperature concerned (up to >900°C.).

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THE friction and wear of metallic surfaces sliding over other surfaces are of great practical importance in affecting the efficiency and life of modern machines. One fundamental feature of both friction and wear, whether in the presence of a lubricating-oil film or not, is the elastic and plastic deformation of those parts of the surface which undergo stress as a result of contacts with the opposite surface. Little evidence of the nature of the surface deformation is obtainable from optical or electron microscopy, which reveal only local areas where metallic welding has occurred across the interface, and long grooves where abrasive particles have ploughed their way along the surface, with occasional evidence of translational slip bands near such regions. X-ray diffraction results also have revealed little more than that a disturbed surface region up to about 10μ thick is caused by such processes as abrading or polishing a crystal surface.[1]

Since the discovery [2, 3] of the existence and nature of a new deformation process in crystals (rotational slip), electron-diffraction results [2, 4, 5] (*cf.* also Germer[6]) have shown that unidirectional abrasion of crystal surfaces, including copper and iron, generally leads to large rotations of parts of the crystal lattice about a well defined axis which is normal to a densely populated net-plane, usually a plane which is parallel or almost parallel to the abrasion direction. The present experiments[7] on iron crystals, using only very light abrasion, not only show further examples of this type, which may correspond to rotational slip, but also bring to light some new modes of surface deformation and associated thermal changes caused by abrasion.

Manuscript received 29th March, 1954.

Dr. Agarwala was formerly a research student in the Department of Chemical Engineering at the Imperial College of Science and Technology, and Dr. Wilman is Lecturer in Applied Physical Chemistry in the same Department.

EXPERIMENTAL DETAILS

Two iron crystals were used in these experiments. They were about 2 mm. thick × 2·5–4 sq. cm. in area, and were cut from strips containing several crystals, prepared by the strain-anneal method. After smoothing on emery paper down to 0000 grade, in the presence of benzene, they were electropolished and examined by electron diffraction as described previously by Evans, Layton, and Wilman,[5] except that 120-kV. electrons were used for the electron diffraction, so as to reduce the background scattering in the photographs from these relatively rough surfaces. The camera length used was in all cases 48 cm.

The main surface of one crystal was of $(h0l)$ type, $8\frac{1}{2}°$ inclined to a (001) plane, and thus near (107). The other crystal had a practically (110) surface, actually inclined at 2° 20′ to (110) about the [110] lattice row. Both crystals yielded, after electropolishing, clear patterns of sharp Kikuchi lines and

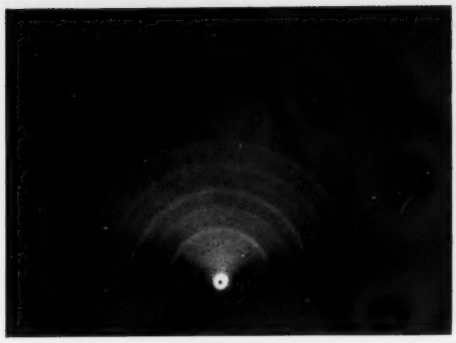


Fig. 1—Electron-diffraction pattern from freshly abraded α-iron crystal; α-iron, random crystals

Fig. 2—Abraded iron crystal etched 1 min.: rings show f.c.c. γ iron

(a)

(b)

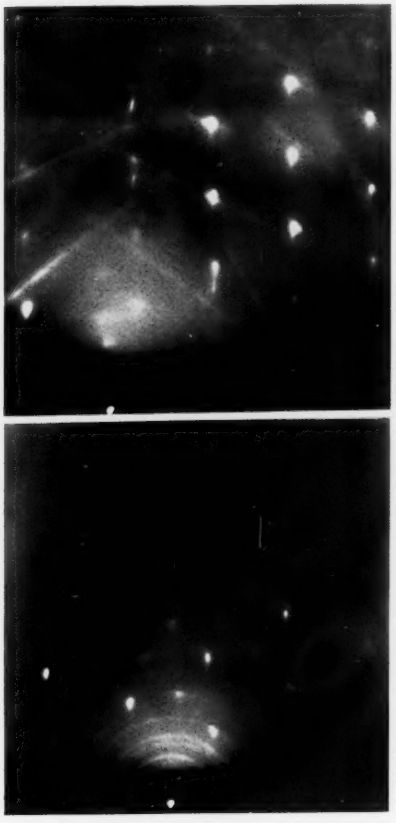
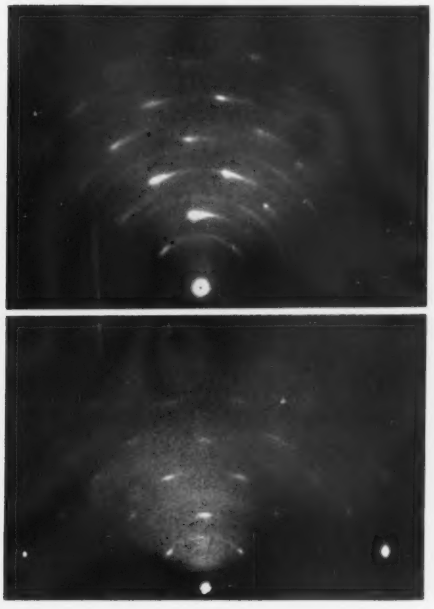


Fig. 3—(*a*) Electropolished α-iron (107) face, beam along '[120]'; (*b*) As *a* after '[210]' abrasion and etch: beam along abrasion direction



(a)

(b)

Fig. 4—The (107) α-iron face abraded '[100]' and etched: beam (*a*) normal and (*b*) parallel to abrasion direction

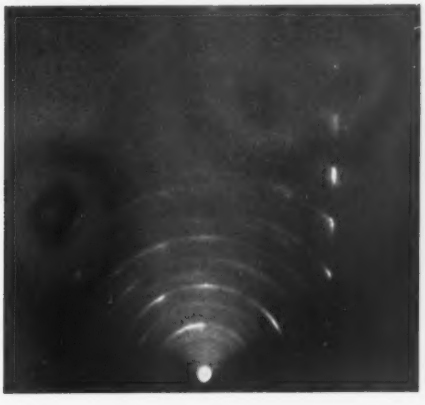


Fig. 5—The (110) α-iron face abraded $[\bar{1}11]$ and etched; beam normal to abrasion (repeat experiment)

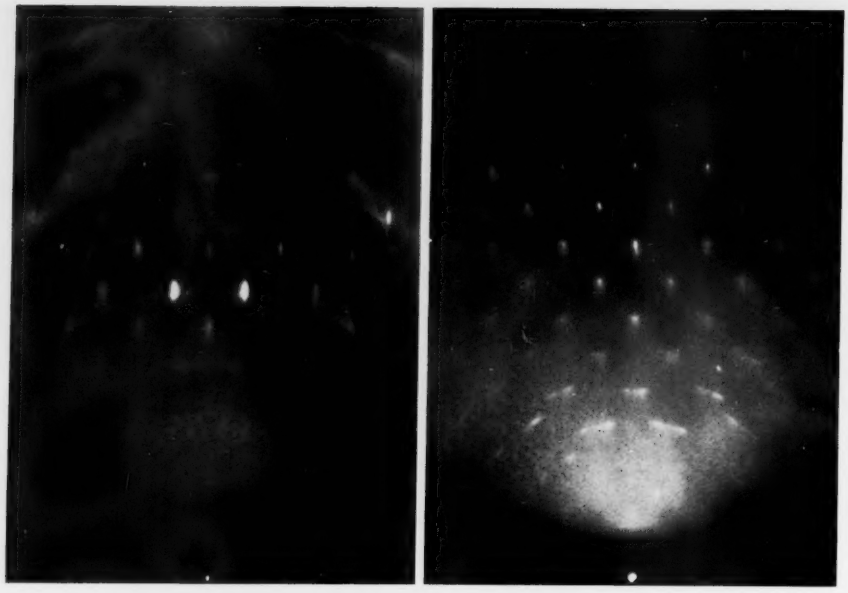


Fig. 6—(*a*) The (107) α-iron face, electropolished, and beam '[100]'; (*b*) As *a* after abrasion '$[\bar{1}00]$' and etch

Fig. 7—The (110) α-iron face abraded $[\bar{1}1\bar{2}]$ and etched: beam (*a*) normal and (*b*) parallel to abrasion

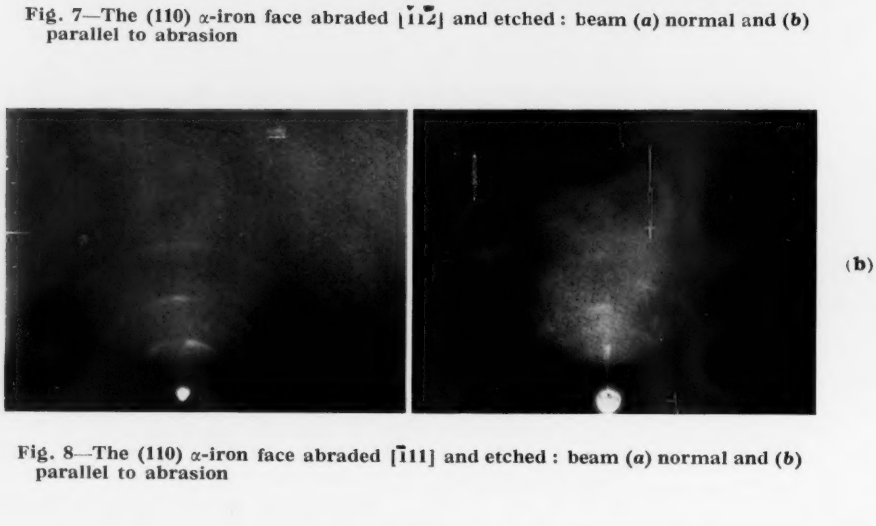


Fig. 8—The (110) α-iron face abraded $[\bar{1}11]$ and etched: beam (*a*) normal and (*b*) parallel to abrasion

elongated spots showing their high perfection and atomically smooth surface.

These smooth crystal faces were abraded by a single continuous stroke 10 in. long, at about 3 in./sec., in any desired direction on 0000 emery paper flooded with benzene. Only light hand pressure was used on the crystal, equivalent to about 50 g./sq. cm. The abraded surfaces were then examined again by electron diffraction after successive periods of etching in 1% picric acid in alcohol, or electropolishing, followed by washing in a jet of acetone, and transference to the electron-diffraction camera under *n*-propyl alcohol to avoid undue oxidation. After electron-diffraction examination the surface was at most stages also examined by optical microscopy using a Vickers projection microscope.

EXPERIMENTAL RESULTS

The abrasion direction is defined as the direction in which the abrasive particles moved relative to the crystal. In all the experiments the surface mainly yielded ring patterns such as Fig. 1 immediately after abrasion, showing randomly disorientated fragments of normal b.c.c. α iron, with lattice dimension $a = 2{\cdot}86$Å. After slight etching (~45 sec.) in a 1% picric acid solution in alcohol, a ring pattern corresponding to a f.c.c. structure with a cube axis $a = 3{\cdot}60 \pm 0{\cdot}01$Å. relative to $a_\alpha = 2{\cdot}86$Å. was obtained, either alone as in Fig. 2, or still with presence of some rings from the b.c.c. α iron. The lattice dimension of the f.c.c. material, together with further evidence described later, shows that this material consists of γ iron. The formation of this γ iron evidently indicates that the temperature of parts of the iron crystal surface had risen to about 900°C. during abrasion.

After further etching (~2 min.) the γ-iron ring pattern usually had disappeared and only rings and arcs due to α iron remained. After total etching of about 4–6 min., most of these randomly disorientated α-iron fragments had been removed, and the patterns then showed some spots corresponding to the main α-iron crystal orientation, a trace of a ring pattern due to γ-Fe_2O_3, and well defined arcs on the α-iron ring positions. The arcs due to the α iron indicated in almost all cases a range of rotations from the initial crystal orientation, about a well defined axis.

The observations after etching or electropolishing down to this region are described below. In all cases the sense of the lattice rotations observed was the same as that of a wheel rolling on the surface along the abrasion direction or nearly so, with its axis normal to a main lattice plane. For convenience in considering the relation between the various azimuths, all the diagrams showing indexing of the diffractions correspond to the negative photographic plate as viewed along the beam direction, whereas the patterns illustrated are positives and thus are reversed from left to right relative to the negative or the index diagram.

Experiments on the (*h*0*l*) Face Inclined at 8½° to (001)

Since the (001) plane was so nearly parallel to the surface, it is most convenient to refer to the azimuths in the surface in terms of the nearest lattice-row directions in the (001) plane.

Abrasion Parallel to the '[$\bar{1}$10]' *Direction*—After etching for 2¼ min., the surface yielded a pattern of clear Kikuchi lines and some strong spots corresponding to the initial crystal orientation, and, in addition, when the beam was normal to the abrasion direction there were strong arcs tailing off in intensity in one direction. With the beam in azimuths on one side or other of the above, these asymmetric arcs became shorter. Thus it is clear that these parts of the lattice had been rotated through 40–50° about an axis which was evidently approximately [110]. The pattern obtained with the beam along the abrasion direction, [$\bar{1}$10], contained many spots corresponding to the initial crystal orientation and many additional spots or short arcs, occurring on nearly vertical layer lines midway between the nearly vertical rows of spots due to the initial crystal orientation. This confirms to a high accuracy that the rotation had indeed occurred about the [110] axis. The additional arcs which correspond to reciprocal-lattice points crossing the Ewald sphere of reflection during the rotation had a spread of about $\pm$ 2° from the mean, round the electron beam direction. A further etching for 1½ min. caused little change in the diffraction patterns, and still further etching merely reduced the angular range of the rotation progressively until the practically undeformed underlying crystal was reached.

These results are closely similar to those found for iron by Evans, Layton, and Wilman [5] (in their case (iii)) and so the photographs are not further illustrated here. In the present crystal, however, the surface was inclined to (001) by 8½° about [010] instead of by 10° about a <110> row.

Abrasion along '[$\bar{2}$10]'—After the re-electropolished surface had been abraded and then etched for 4 min., it yielded electron-diffraction patterns of well defined Kikuchi lines and some strong spots corresponding to the initial crystal orientation; and also, when the beam was normal to the abrasion direction, long arcs were present, tailing off in one direction from the normal spots. The angular range of the stronger part of these arcs round the undeflected beam was about 20–25°, and the arcs extended with appreciable intensity up to 40–50°. When the beam grazed the crystal surface in azimuths on one side or other of this, the asymmetric arcs became progressively shorter. Parts of the lattice near the surface had thus rotated by up to 50° about an axis evidently approximately parallel to the [120] lattice row, which was practically normal to the abrasion direction.

In agreement with this conclusion, the diffraction arcs were shortest, and showed no asymmetric tailing off in intensity to one side or the other, when the electron beam was along the abrasion direction, '[$\bar{2}$10],' as Fig. 3*b* shows. The pattern at this azimuth shows clearly (to within 2–3°) that the rotation axis was the [120] row in the (001) plane, not the direction lying in the crystal surface normal to the abrasion direction. In Fig. 3*b* the strong spots correspond to parts of the surface which had still the initial crystal orientation. These spots lie in a centred $\sqrt{5}$-rectangle pattern with the long side

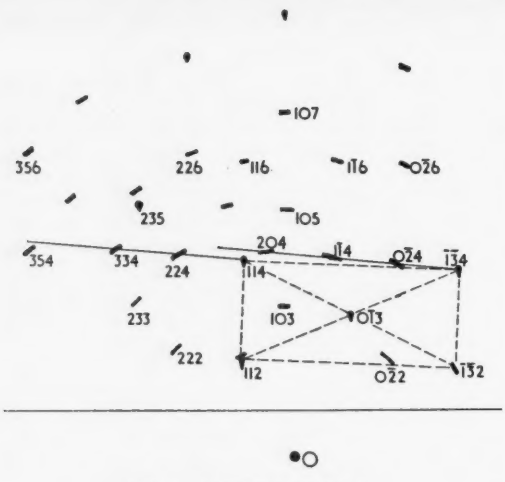


Fig. 9—Indexing of Fig. 3*b*

inclined at about 5° to the (horizontal) shadow edge. Figure 3*a* shows such a pattern from the initial electropolished surface, and the elongated spots and sharp Kikuchi lines also indicate the high smoothness and perfection of the crystal initially at the surface. The indexing of the normal spots and the very short arcs in Fig. 3*b* is shown in Fig. 9, which was constructed from a reciprocal lattice plan on (001), in the way described by Evans, Layton, and Wilman.[5] Note, for example, that the $0\bar{2}4$, $1\bar{1}4$, and 204 diffractions occur at points distant 0·3, 0·6, and 0·9 times the long side of the $\sqrt{5}$ rectangle from the strong $\bar{1}34$ spot, in agreement with the conclusion that the rotation axis was [120]. The extent of arcing of these additional diffractions in Fig. 3*b* is only about $\pm$ 1½–2° about the abrasion direction, in the case of those diffractions which occur as a result of the lattice rotation attaining 12° from the initial orientation, about the [120] axis normal to the abrasion direction.

On etching the crystal further in stages, the electron-diffraction patterns were of similar type to those just described, but indicated a progressively smaller range of rotation of the lattice at deeper levels. The crystal was therefore electropolished far enough to remove the disturbed layer caused by the abrasion, and the electron-diffraction pattern was then similar to Fig. 3*a*, showing the surface to be highly perfect and atomically smooth. The experiment was then repeated (*i.e.*, the crystal was abraded along '[$\bar{2}$10]') and again the results were closely similar to those described.

Abrasion along '[100]'—The crystal was next electropolished far enough to remove effectively all the surface region disturbed during the previous abrasion experiment, and the electron-diffraction patterns of Kikuchi lines and spots showed the surface to be again highly perfect and smooth. After a single abrasion stroke along [100], followed by etching for 45 sec., the patterns of Figs. 4*a* and *b* were obtained with the beam respectively perpendicular to and parallel to the abrasion direction. Figure 4*a* consists almost entirely of the normal square spot pattern associated with the initial crystal orientation, rotated round the undeflected beam through about 50–60° arc, the diffraction intensity tailing off continuously in one direction along each arc. Thus, parts of the lattice had been rotated by about 50–60° about an axis parallel to or very close to the beam direction, *i.e.*, [010].

Figure 4*b*, obtained with the beam along the abrasion direction, shows short symmetrical arcs arranged in a square pattern. These arcs correspond to the square array of lattice points in the (100)* plane through the origin in the reciprocal lattice. The points corresponding to these arcs cross the Ewald sphere of reflection almost simultaneously after a rotation of 8½° from the initial crystal orientation. In comparison with this rotation of 8½° in the main direction, about [010], it is seen that the slight arcing of these diffractions shows a spread of about $\pm$ 3° of rotation about the abrasion direction, *i.e.*, '[100].' Other fainter additional diffraction arcs are also present, corresponding to parts of the lattice rotated through angles larger than 8½°.

Etching for a further 1½ min. reduced the arcing by about half, and a further etching for 45 sec. reduced the range of rotation almost to zero.

The crystal was then again electropolished to prepare a practically smooth single-crystal surface, and the abrasion along [100] was again repeated, with practically identical results.

Abrasion along '[$\bar{1}$00]'—A further abrasion was carried out on the crystal after freshly resurfacing it by electropolishing, but this time the abrasive motion was along '[$\bar{1}$00],' *i.e.*, opposite to that described in the previous sub-section. After etching for several minutes, the surface yielded a pattern similar to Fig. 4*a*, when the electron beam was perpendicular to the abrasion direction. The long arcs tailing off in one direction show that parts of the lattice had rotated by up to 40–50° about an axis roughly in the surface and normal to the abrasion direction. When the beam was along the abrasion direction, however, the patterns consisted mainly of rings.

The surface was examined after various stages of etching up to 12 min., but the electron-diffraction patterns obtained with the beam along the abrasion direction still did not show any well marked vertical layer lines, whilst with the beam perpendicular to the abrasion direction the asymmetric arcs in the patterns became progressively shorter down to only about 10°. On electropolishing the surface for a further 45 sec., the asymmetric arcs were reduced to only about 5° range extending from the strong spots which corresponded to those parts of the surface still having the initial crystal orientation. Figure 6*b*, however, was then obtained when the beam was in the direction [100], opposite to the abrasion direction.†

In the upper part of Fig. 6*b*, in part of a circular zone round the point where the [100] cube edge meets the plate, there are sharp strong spots elongated vertically normal to the shadow edge. These spots

†Figures 6*a* and *b* have been reproduced, with brief comment, elsewhere (Wilman, *op. cit.*,[1] p. 314).

correspond to parts of the surface still having the initial crystal orientation, as may be seen by comparison with Fig. 6*a*, which was obtained from the crystal before the abrasion was carried out. In the lower parts of Fig. 6*b* there are additional short arcs which at first sight suggest the extension of the initial spot pattern as usual along the vertical layer lines, as seemed to be indicated from the tailed arcs obtained at the perpendicular azimuth. However, these additional diffractions in Fig. 6*b* are not single short arcs, but pairs of spots joined together by a much fainter continuous arc. Both spots of each pair lie at the same radial distance from the undeflected-beam spot, but are displaced approximately equally to one side or the other from the vertical rows of spots of the initial crystal pattern.

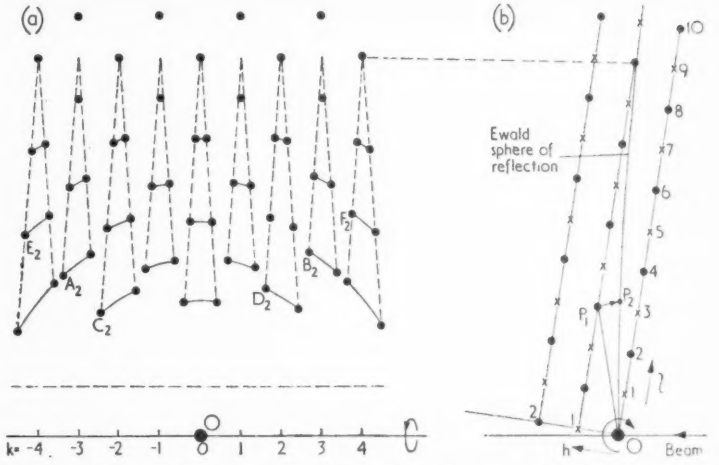


Fig. 10—Diagram for indexing of Fig. 6*b* : (*a*) spot positions ; (*b*) side view (along rotation axis [010]) of part of reciprocal lattice in initial orientation, × λL, to same scale

These features, together with the regular increase in the distance apart of the spots of these pairs towards the lower part of the pattern, show that in this case some parts of the lattice had rotated about one axis, and others about another axis, these two axes being symmetrically disposed relative to the (010) plane, but inclined away from the [010] axis. The spots which occurred in Fig. 6*b* as a result of these rotations show that the extent of the rotation still present at the crystal surface at this stage of etching was only about 5°, in agreement with the extent of the asymmetric arcing in the perpendicular azimuth.

The directions of these rotation axes were calculated as follows, by the use of Ewald's reciprocal-lattice construction. First the positions (*see* Fig. 10*a*) of the strong spots corresponding to the initial orientation were used in the usual way to define the location of the Ewald sphere of reflection relative to the reciprocal lattice, as shown in the section on the (010)* plane in Fig. 10*b*. Next the co-ordinates [[*hkl*]]* (relative to the reciprocal axes a^*, b^*, c^*) of points such as P_2 corresponding to the observed additional spots, were found relative to the axes a^*, b^*, and c^* of the initial reciprocal lattice. The h_2 and l_2 co-ordinates, relative to a^* and c^*, were obtained sufficiently accurately from a diagram such as Fig. 10*b*, drawn to a scale such that the sphere radius was 480 cm. (*i.e.*, the camera length × 10). Then k_2 was estimated directly from the spot position in Fig. 6*b* relative to the nearly vertical rows of normal spots in the upper region of the pattern. The direction of these rows of spots was defined most accurately from the row which passes through the undeflected-beam spot. Each point such as P_2 on the Ewald sphere of reflection in Fig. 10*b*, corresponding to an additional spot in the pattern, is brought into this position P_2 by the rotation from the initial point P_1, whose co-ordinates are $[[h_1\ k_1\ l_1]]^*$ relative to the initial reciprocal-lattice axes. The line joining P_1 and P_2 has thus the indices $[uvw]^*$ given by $[h_2–h_1,\ k_2–k_1,\ l_2–l_1]^*$ and is perpendicular to the rotation axis ; thus the indices $[UVW]^*$ of the rotation axis are related to $[uvw]^*$ by the relation

$$uU + vV + wW = 0 \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (1)$$

Table I shows the co-ordinates [[*hkl*]]* of the initial reciprocal-lattice points A_1, B_1, C_1, etc., and the points A_2, B_2, C_2, etc., where, during the rotation, they cross the Ewald sphere of reflection as shown in Fig. 10*b*. The points A_2, B_2, etc., were all points associated with the same rotation axis of the symmetrical pair. Table II shows the indices $[uvw]^*$ of the lines A_1A_2, B_1B_2, etc.

By modifying the co-ordinates $[uvw]^*$ in Table II to take the form $[u'v'1]^*$, suitable pairs of equations of the type (1) could be combined to give directly the ratio U/V. The nine most accurate estimates of

Table I

CO-ORDINATES [[*hkl*]]*

Relative to the reciprocal lattice of the initial crystal, of initial lattice points A_1, B_1, C_1, etc., and of the points A_2, B_2, C_2, etc., where, during the rotation, they cross the sphere of reflection, in the case of abrasion of (107) along '[$\bar{1}$00].'

	A_1	A_2	B_1	B_2	C_1	C_2
h	−1	−0·49	−1	−0·57	−1	−0·38
k	−3	−3·40	3	2·70	−2	−2·43
l	−4	−3·65	−4	−4·35	−3	−2·77
	D_1	D_2	E_1	E_2	F_1	F_2
h	−1	−0·44	−1	−0·60	−1	−0·66
k	2	1·71	−4	−4·35	4	3·81
l	−3	−3·31	−5	−4·67	−5	−5·37

Table II

INDICES [*uvw*]* OF THE LINES JOINING PAIRS OF POINTS SUCH AS A_2 AND A_1 OF TABLE I

	A_2–A_1	B_2–B_1	C_2–C_1	D_2–D_1	E_2–E_1	F_2–F_1
u	0·51	0·43	0·62	0·56	0·40	0·34
v	−0·40	−0·30	−0·43	−0·29	−0·35	−0·19
w	0·35	−0·35	0·23	−0·31	0·33	−0·37

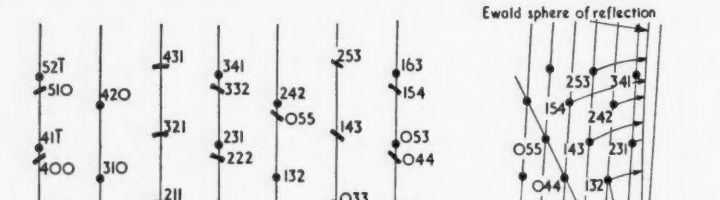


Fig. 11—Indexing of Fig. 7*b*: (*a*) diffraction positions and indices; (*b*) side view (along rotation axis) of part of reciprocal lattice, × λL, to same scale

V/U were 1·35, 1·57, 1·41, 1·44, 1·61, 1·50, 1·31, 1·51, and 1·31, and the mean of these gave $U/V = 1 : 1{\cdot}44$. Similarly, the more accurate estimates of W/U were 0·08, 0·33, 0·19, 0·00, 0·29, 0·15, 0·14, 0·38, and 0·23, giving a mean of $U/W = 1 : 0{\cdot}20$. Thus the combined measurements for the six diffraction spots *A–F* gave as a combined mean estimate $U/V/W = 1 : 1{\cdot}44 : 0{\cdot}20$. Within the accuracy obtained, it seems probable that this axis of rotation corresponds to the [451] axis of flexure of ($\bar{2}$13) slip lamellæ, since [451] is perpendicular to the slip direction [1$\bar{1}$1] in the ($\bar{2}$13) plane. The other axis of rotation, symmetrical across (010), is the [4$\bar{5}$1] direction, which lies in the ($\bar{2}\bar{1}$3) plane and is normal to [111].

Experiments on the Approximately (110) Face

The crystal face, carefully surfaced by electropolishing, yielded clear patterns of strong Kikuchi lines and spots elongated normal to the shadow edge, showing the high perfection of the crystal and the atomically smooth nature of the surface. In the optical microscope the surface appeared practically featureless.

Abrasion along [$\bar{1}$11]—The surface was first abraded along [$\bar{1}$11] and after light etching yielded the usual electron-diffraction pattern (Fig. 5) of long arcs tailing off in one direction from the spots of the normal pattern, when the beam was normal to the abrasion direction. The rotation of the surface regions of the lattice had thus evidently been about [1$\bar{1}$2] or some neighbouring direction. However, when the beam was along the abrasion direction, only ring patterns were obtained, showing randomly disorientated fragments, and no clear groups of arcs along vertical lines were observed even after much further etching, down to the practically undisturbed region of the crystal. Although the rotation axis was thus not identifiable more closely, it corresponded to that in the sub-section below dealing with abrasion along [$\bar{1}$11].

After the crystal had been resurfaced by electropolishing for 5–6 min. the following experiment was carried out.

Abrasion along a Direction 8° *from* [$\bar{1}$1$\bar{2}$] (*on the side towards* [00$\bar{1}$])—Here also, long arcs were observed tailing off in one direction from the spots of the normal pattern, when the beam was normal to the abrasion direction. The fragments were therefore rotated by up to 45° from the initial orientation, about an axis along or near [$\bar{1}$11]. Even after repeated etching and slight electropolishing, however, the pattern obtained with the beam along [$\bar{1}$1$\bar{2}$] (*i.e.*, 8° away from the abrasion direction) consisted mainly of rings from disorientated iron fragments; thus closer specification of the axis of rotation was impossible. After further electropolishing for a few seconds the crystal yielded at all azimuths clear patterns of spots due to γ iron orientated in a definite way relative to the underlying α iron.

Similar patterns were also observed in the experiments described in the next two sub-sections. They are dealt with briefly on p. 130 and are discussed in detail elsewhere.[8]

Abrasion along [$\bar{1}$1$\bar{2}$]—The crystal, resurfaced by electropolishing for 5–6 min., was abraded in one stroke along [$\bar{1}$1$\bar{2}$] and then yielded only ring patterns showing randomly disorientated iron fragments. After etching for 4 min., arcs were observed tailing off in one direction from the normal spots, when the beam was normal to the abrasion direction; but rings were still obtained when the beam was along the abrasion direction. The surface was therefore electropolished for 5 sec. and then yielded the patterns shown in Figs. 7*a* and *b* when the beam grazed one part of the surface respectively normal to and parallel to the abrasion direction. A different region of the sursurface, however, now yielded, at various azimuths, patterns consisting mainly of strong sharp spots, with faint Kikuchi lines also present. These patterns, associated with the presence of γ iron, are described

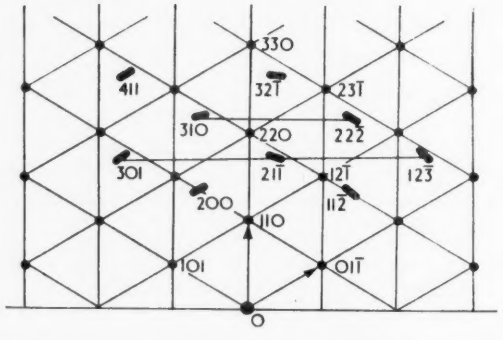


Fig. 12—Theoretical diffraction spot and arc positions for a rotation about [$\bar{1}$1$\bar{2}$] from the initial setting with (110) parallel to the beam

elsewhere.[8] Visual inspection showed that the surface had not been uniformly abraded, presumably because, although highly smooth, it was not perfectly flat. Thus the subsequent etching and electropolishing had removed almost completely the thin surface layer of disorientated material from the less abraded regions, although it had removed only partially the thicker layer from the more heavily abraded parts.

Figure 7*a*, evidently from the more heavily abraded region, was obtained when the beam was along [1$\bar{1}\bar{1}$], *i.e.*, normal to the abrasion direction. The normal hexagon spot pattern is present, together with arcs extending in one direction from these spots round the ring positions, the strongest parts of each arc subtending at the undeflected-beam spot an angle of 30°, although the arcs extended more weakly up to as much as 60°. This indicates that the axis about which the fragments had been rotated from the initial orientation was approximately [1$\bar{1}\bar{1}$], *i.e.*, normal to the (1$\bar{1}\bar{1}$) planes. Figure 7*b*, obtained from the same region with the beam along [$\bar{1}$1$\bar{2}$], shows spots and very short arcs lying on the corresponding roughly vertical layer lines. A detailed analysis of the pattern, leading to the indexing shown in Fig. 11, confirms that the rotation axis was indeed [1$\bar{1}\bar{1}$] within an accuracy of estimation of 2–3°.

By comparing Fig. 7*b* with Fig. 11, it is seen that the spread of the diffractions is small relative to the angle of rotation (from the initial lattice) required for the diffractions to occur. For example, the 321 diffraction has a spread of only about ± 3° from the mean, although it occurs as a result of a lattice rotation of roughly 16° round the [1$\bar{1}\bar{1}$] rotation axis.

Abrasion along [$\bar{1}$11]—The crystal was next electropolished for 5–6 min. to remove all traces of the previous abrasion and to prepare again a smooth, optically almost featureless (110) surface. It was then abraded along [$\bar{1}$11]. Etching for 2 min., followed by electropolishing for 2 sec., removed the surface layer of randomly disorientated α and γ iron and exposed a region which yielded Figs. 8*a* and *b*, when the electron beam was respectively perpendicular to and in the same direction as the abrasion direction.

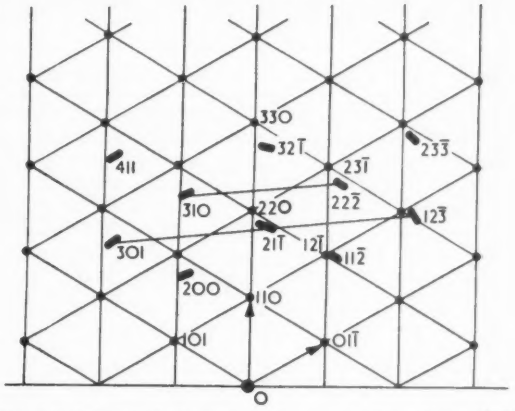


Fig. 14—Theoretical pattern for a rotation about [$\bar{3}$3$\bar{4}$]

Figure 8*a* shows as usual long arcs tailing off in one direction from the normal diffraction spot positions associated with the initial crystal. Parts of the lattice had thus evidently rotated about an axis which was not far from the beam direction [$\bar{1}$1$\bar{2}$]. Figure 8*b* confirms this conclusion by the positions of the arcs and the fact that they are short arcs and not tailing off asymmetrically. Here, however, it is clear that the rotation had not been about [$\bar{1}$1$\bar{2}$], which would have caused the main diffraction arcs to be in the positions shown in Fig. 12, constructed from the plan shown in Fig. 13. For example, the 301, 21$\bar{1}$, and 12$\bar{3}$ arcs would then lie on a horizontal line (normal to the 110 radius vector) in Fig. 12, instead of a line observed in Fig. 8*b* to be inclined at 3½° to the corresponding direction ; and these arcs would be displaced too far to the right. On the other hand, the rotation cannot have been about the [$\bar{1}$1$\bar{1}$] row which was in the surface (110) plane but at 21° to [$\bar{1}$1$\bar{2}$], because this would have resulted in the 21$\bar{1}$ and 32$\bar{1}$ arcs being centred on the plane of incidence, which is not the case.

The method used on p. 126, in the sub-section " Abrasion along [$\bar{1}$00]," was therefore applied to estimate the rotation axis directly from the observed arc positions.

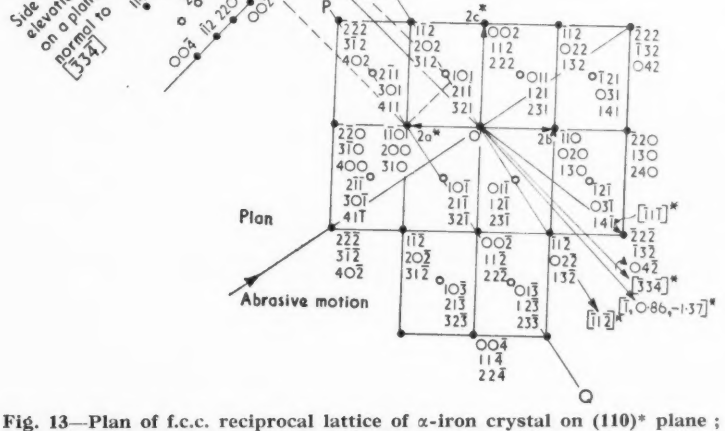


Fig. 13—Plan of f.c.c. reciprocal lattice of α-iron crystal on (110)* plane ; and side elevation for construction of pattern expected from rotation about [$\bar{1}$1$\bar{2}$], etc.

Table III

CO-ORDINATES [[*hkl*]]*

Relative to the reciprocal lattice of the initial crystal, of initial points A_1, B_1, C_1, etc., and of points A_2, B_2, C_2, etc., where during the rotation, they cross the sphere of reflection, in the case of abrasion of (110) along [$\bar{1}11$].

	A_1	A_2	B_1	B_2	C_1	C_2	D_1
h	2	1·64	3	2·52	1	0·75	2
k	1	1·81	2	2·76	1	1·97	2
l	−1	−0·17	−1	−0·23	−2	−1·22	−2

	D_2	E_1	E_2	F_1	F_2	G_1	G_2
h	1·64	1	0·76	3	2·50	3	2·52
k	2·83	2	2·94	0	0·70	1	1·65
l	−1·19	−3	−2·18	1	1·80	0	0·87

Table III shows the indices of reciprocal-lattice points lying in the $(1\bar{1}\bar{1})$* plane next to that through the origin, seen in the plan in Fig. 13 as the line *PQ*. The indices (relative to the axes of the initial lattice) are also given for the points where these lattice points cross the Ewald sphere of reflection during the rotation, as calculated from the observed arc positions relative to the faint sharp spots from the initial crystal orientation (marked by small filled-in circles in Figs. 12 and 14). The differences between the two sets of indices, given in Table IV, represent the indices of lines which are normal to the rotation axis [*UVW*]*, and led as before, when taking the least inaccurate combinations, to the *V*/*U* ratios −0·37, −1·07, −0·77, −0·43, −0·52, −2·13, −1·12, −0·37, −0·83, −1·27, −0·65, −1·12, and −0·51, and to the *W*/*U* ratios 0·86, 1·72, 1·39, 0·89, 1·03, 2·65, 1·71, 0·83, 1·28, 1·77, 1·03, 1·77, and 0·92.

The mean estimate thus gives $UVW = -1 : +0{\cdot}86 : -1{\cdot}37$, defining [*UVW*] to an accuracy of 2–3°. This axis is shown in the plan in Fig. 13 and is inclined downward about 4° from the horizontal plane (110) on which the plan is made. A diagram was constructed, showing the pattern of arcs which would result from rotation of the initial crystal lattice about this axis and this still showed some appreciable discrepancies with the observed pattern, Fig. 8*b*. Much better agreement was found with Fig. 14, where the positions of the main arcs are shown resulting from a rotation of roughly 30° about an axis [$\bar{3}3\bar{4}$], which lies in the (110) plane near the above-estimated direction of the rotation axis. The main discrepancy still present is that the rows of arcs such as that on which lie 301, $21\bar{1}$, $12\bar{3}$ are at 5° to the horizontal, (*i.e.*, the normal to the 110 radius vector in Fig. 14) instead of the $3\frac{1}{2}$° observed, and this suggests that the rotation axis is at about this azimuth in the (110) plane but inclined down about 1° (and thus slightly more inclined away from the actual crystal surface than was the (110) plane).

The above rotation thus accounts for the symmetrical arcs in Fig. 8*b* and also for the arcs in Fig. 8*a* which tail off in one direction from the spots associated with the initial crystal orientation. There are, however, other prominent arcs still not explained in Fig. 8*a*, *e.g.*, on the 110, 211, 321, and 310 ring positions. At first sight it seemed necessary to postulate that parts of the lattice had rotated about the axis near [$\bar{1}10$], which would account quite well for the *locations* of the arcs concerned on these ring positions. It would also account largely for their relative intensities except, however, for the marked absence of a $2\bar{2}\bar{2}$ diffraction centred on or near the plane of incidence while the $1\bar{1}\bar{2}$ arc was moderately strong.

The true explanation, however, for the occurrence of all these additional arcs is evidently that they result also from the [$\bar{3}3\bar{4}$] rotation just described, but appear only because the surface projections through which the electron beam can pass, and which therefore contribute to the pattern, are thin (*i.e.*, of limited extent) in the beam direction. This corresponds, in terms of Ewald's reciprocal-lattice construction, to an extension in or near the beam direction, of the reciprocal-lattice 'intensity regions' surrounding the lattice points, so that these extensions cross the Ewald sphere of reflection even though the lattice points do not lie on the sphere. In the present case such extensions, together with the established [$\bar{3}3\bar{4}$] rotation, would indeed give rise to the observed arcs. The corresponding and essentially closely similar pattern Fig. 5 (*see* p. 128), obtained with the beam normal to the same [$\bar{1}11$] abrasion direction but in the opposite sense, shows more obviously the correctness of this interpretation.

Evidence of an Orientated Gamma-Iron Layer

The patterns of sharp spots referred to on p. 128, which were obtained at a late stage of etching (after abrasion), when the rotationally disorientated layer of α iron had been removed, have now been fully elucidated. The results are described in detail elsewhere.[8] Here it may be stated that these patterns, at any given azimuth, are independent of the direction of abrasion, and correspond to f.c.c. γ iron strongly orientated relative to the underlying practically undistorted main α-iron crystal.

The observed orientations of the γ iron relative to the α iron are different from those observed by Kurdjumow and Sachs,[9] and others quoted by Barrett,[10] and are such that a {110}γ plane is parallel to a $(1\bar{1}1)$ or $(1\bar{1}\bar{1})\alpha$ plane (normal to the (110) surface of the crystal), with a perpendicular {111}γ plane inclined at 5° to the (110) surface of the α-iron crystal. This is more simply expressed as of the type having a (001)γ plane parallel to a $(01\bar{1})\alpha$ plane, with [$\bar{1}10$]γ parallel to [$\bar{1}11$]α. These main γ-iron lattices were extensively twinned on {111}γ.

At this stage of etching, patterns were also obtained from parts of the crystal surface, clearly due to the main α-iron crystal but containing groups of four spots instead of each spot of the normal pattern.† The spot positions correspond to extensions of the reciprocal-lattice intensity regions along ⟨110⟩α and

† One of these patterns is reproduced elsewhere (Wilman, *op. cit.*,[1] p. 315, Fig. D10*b*).

Table IV

INDICES [*uvw*]* OF THE LINES JOINING PAIRS OF POINTS SUCH AS A_2 AND A_1 OF TABLE III

	A_2–A_1	B_2–B_1	C_2–C_1	D_2–D_1	E_2–E_1	F_2–F_1	G_2–G_1
u	0·37	0·48	0·26	0·36	0·24	0·50	0·49
v	−0·81	−0·76	−0·97	−0·83	−0·94	−0·70	−0·65
w	−0·83	−0·77	−0·78	−0·81	−0·82	−0·80	−0·87

thus are clear evidence of {110} faces on the α-iron surface. Since no such patterns were ever obtained by the etching or electropolishing alone, these patterns indicate, in agreement with the above orientation, that the boundary faces between the γ and α lattices are of {110}α, *i.e.*, {001}γ type.

These relative orientations, and the interface form, lead to the conclusion that this γ iron is formed by a homogeneous shear of {211}α planes past each other along ⟨111⟩α (*i.e.*, ⟨110⟩γ), as in normal twinning, accompanied by a small homogeneous perpendicular compression.

DISCUSSION

The results of these experiments on the (*h*0*l*) face inclined at 8½° to (001), and on the approximately (110) face, provide much information additional to that described by Evans, Layton, and Wilman,[5] but in the present experiments a very much lighter abrasion by a single 10-in. stroke was used. In most cases large rotations occurred about a well defined axis normal to a densely populated plane. Thus, on the (107) face, abrasion along the '[$\bar{1}$10]' '[$\bar{2}$10],' and '[100]' azimuths, respectively, led to rotations about the lattice directions [110], [120], and [010], respectively. On the nearly (110) face, abrasion along [$\bar{1}1\bar{2}$] and also along a direction 8° from this, on the side towards [00$\bar{1}$], gave rise to a rotation about [1$\bar{1}\bar{1}$].

Of these results only that of abrasion along '[$\bar{2}$10]' on (107) corresponds to an abrasion direction which is not parallel to a plane of symmetry of the lattice, and it is noteworthy that in this case too, the observed rotation axis was normal to a densely populated plane (120) which lay parallel to the abrasion direction. It is difficult to account for this rotation by any translational-slip system, even if it were possible to account for the other cases as double translational slip on two systems mirror-symmetrical to each other across the symmetry plane. However, all these rotations could correspond quite simply to a fragmentation involving formation of coherent lamellæ in a rotational-slip relationship to each other and to the main crystal (*cf.* Wilman [2, 3] and Evans, Layton, and Wilman[5]).

Besides these relatively simple rotation axes, however, there were two less simple cases observed: (i) abrasion on (107) along '[$\bar{1}$00]' (*i.e.*, [$\bar{7}$01]) gave rise to obscure results near the surface, but to rotations of (presumably) some parts of the sub-surface region about an axis along or close to [451] whilst others had rotated about the axis [4$\bar{5}$1] mirror-symmetrical to [451] across the (010) plane; (ii) abrasion on (110) along [$\bar{1}$11] caused a large rotation about [$\bar{3}3\bar{4}$]. The [451] rotations could correspond to flexure of ($\bar{2}$13) slip lamellæ slipping along [1$\bar{1}$1], but the rotation about [$\bar{3}3\bar{4}$] could not represent flexural translational slip alone, in any of the types of slip along ⟨111⟩ previously observed in iron, because [$\bar{3}3\bar{4}$] is not perpendicular to any ⟨1$\bar{1}$1⟩ row of the lattice. The [$\bar{3}3\bar{4}$] rotation could possibly represent a combination of initial rotational slip (on ($\bar{1}1\bar{1}$) or ($\bar{1}1\bar{2}$) ?) and subsequent translational slip within the lamellæ, or might be a result of multiple translational slip alone.

One factor complicating the interpretation of the above results in terms of slip processes is the inhomogeneous nature of the deformation, corresponding to the abrasive particles being pressed into the surface, pushing away parts of the lattice in front of them, and leaving behind more or less deep furrows. Attempts were made to observe the nature of the initial deformation caused by pressure (similar to that used above) of the electropolished faces against the abrasive particles of the 0000 emery paper, without any tangential motion, but it has not been possible so far to detect by electron diffraction, any clear signs of much deformation, and optical microphotographs only showed sparsely distributed minute dots representing shallow indentations caused by a few particles.

Another factor at present difficult to allow for is the undoubtedly large rise in temperature of the surface layers during the abrasion, as is well shown by the above evidence of the formation of γ iron. However, irrespective of the direction of abrasion, the γ iron in the lower strata only occurred in orientations which corresponded to α to γ transformations involving slip along those ⟨111⟩α rows which lay parallel to the initial (110)α surface of the crystal. This suggests that at least at these levels below the original surface the effect of the abrasion was mainly to cause some translational motion of {211}α planes over each other along these particular usual slip directions, ⟨111⟩α. The rise in temperature (to >900°C.) apparently caused the shear merely to bring about the formation of the γ structure, which remained stable during the rapid cooling after cessation of the abrasion. From this indication that the preferred slip direction is still along ⟨111⟩α (= ⟨110⟩γ) at these temperatures, it seems unlikely that the above rotation about [$\bar{3}3\bar{4}$] can be due to a flexural translational slip.

Acknowledgments

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Ingot Heat Conservation

COOLING OF 15-ton INGOTS BETWEEN TEEMING AND STRIPPING

By L. H. W. Savage, M.Sc., F.R.I.C. and M. D. Ashton, B.Sc.,

SYNOPSIS

The rate of heat loss from the mid-height section of a 15-ton rimming-steel ingot has been determined for 12 casts. It can be represented as a function of time from start of teeming t and initial mould temperature θ by the regression equation:

$$H_s/\sqrt{t} = a + b\theta + ct + dt^2$$

or with slightly greater error by $H_s/\sqrt{t} = k$, where H_s is the heat loss from the steel, and a, b, c, d, and k are constants the values of which have been determined. Increase in initial mould temperature causes a decrease in the rate of heat loss. It is possible to determine initial mould temperature from the time which has elapsed since the mould was stripped.

By making certain simplifying assumptions, the average thickness of solidified shell S can be expressed by the equation:

$$S = K\sqrt{t}.$$

From a limited number of measurements at points other than on the mid-height plane, the average heat loss per unit mass of the ingot was calculated to be about 25% greater than that at the mid-height section.

From the results it is possible to recommend a reduction in standing time between teeming and stripping, which will result in a saving of fuel in soaking pits. 1054

A SERIES of investigations has been carried out by the Fuel Technology Section of B.I.S.R.A. on the cooling of rimming-steel ingots between teeming and stripping. The purpose of this work is to learn more about the thermal condition of ingots about to be charged to soaking pits and the influence of time from teem to strip and other variables upon this condition. From the results it is hoped to effect economy of fuel in soaking pits by charging ingots to them which have the maximum heat content compatible with safe handling. The resulting reduction in time from teem-to-strip also leads to increased mould life.

In a previous paper[1] it was shown that for 8-ton ingots, the heat loss from the steel was a function of time and initial mould temperature, other variables such as initial steel temperature and mould position being of low significance. The work has since been extended to 15-ton ingots cooling in bottle-top moulds at a works where bogie casting is the practice, and a series of 12 casts has been studied using a similar technique to that used before.

The first purpose of this paper is to present the results of this series of trials after the application to them of techniques similar to those applied to the results for 8-ton ingots. The second is to present the results obtained by an extension of the field of measurements to other points in the mould. The third is to outline the requirements for a further series of trials, based upon overcoming the shortcomings of this recent series, and the final purpose is to consider whether the results of this investigation can be applied to obtain earlier stripping of ingots.

List of Symbols

H_s	Heat loss from steel, cal./g. or therms/ton
S	Distance solidified from the steel surface, cm. or in.
t	Time from start of teeming, min.
t'	Time from stripping, hr.
θ	Initial mould temperature, ° C.
θ_m	Mould temperature, ° C.
σ	Standard deviation.

PROCEDURE

As in previous work, measurements on the mould were confined mainly to the mid-height section of the mould into which ten plugs each containing four thermocouples were inserted. A further five plugs were inserted in other parts of the mould and 12 thermocouples, distributed between five plugs, were inserted in one section of the bottom plate. The arrangement of thermocouple plugs in the mould and bottom plate is shown in Figs. 1 and 2 and details of thermocouple plugs are given in Figs. 3*a* and 3*b*.

Temperatures in the mould were recorded by a high-speed recorder/switch-box combination.[2] This combination could record only 44 thermocouples and 4 ' markers ' in each trial, and a means of selecting any 11 out of the 15 plugs in the mould was provided at the recorder end of the system, the other four plugs remaining unrecorded. The cycle of 44 points and 4 marker points took 1–1½ min. Temperatures in the bottom plate were recorded on a 12-point medium-speed recorder.

The temperature differential between the outer pair of thermocouples in each plug was recorded intermittently on a high-speed single-point recorder/switchbox combination. This recorder was also used to follow the temperature rise of one of the innermost thermocouples during the first 10 min. after teeming had commenced.

The recorders were housed in a shelter on a bogie at the front of the train, whilst the mould could

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occupy any one of 12 positions in the train (Fig. 4). The two were connected by a cable terminating in multi-pin plugs and sockets which was fastened along the train to the side of the bogies. This arrangement is illustrated in Fig. 5. Power for the recording instruments was obtained from a 110-V. mains supply when the train was stationary in the casting bay or mould-preparation bay and from a 60-V. battery supply with an A.C. generator at all other times. The instrumentation is described more fully in Appendix I.

Temperatures were recorded during 12 trials from the time when teeming commenced into the test mould until just before stripping. They were also measured at intervals after stripping to establish the rate of cooling of the empty mould. The position of the mould and bottom plate in the train was varied during the trials and variations in initial mould and steel temperatures resulted from works practice. Table I summarizes conditions for the test mould.

RESULTS

Heat Loss from Mid-height Section of Steel

Calculation—For the mid-height section the heat gain by the mould was calculated by the method described by Savage and Fowler,[1] which assumed negligible heat transfer perpendicular to the mid-height plane. The surface heat loss was estimated from measurements of surface temperature gradients

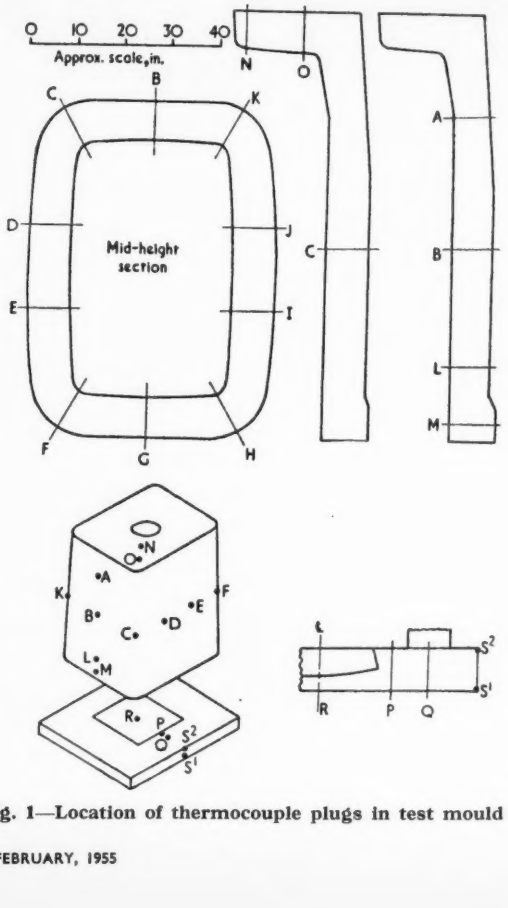


Fig. 1—Location of thermocouple plugs in test mould

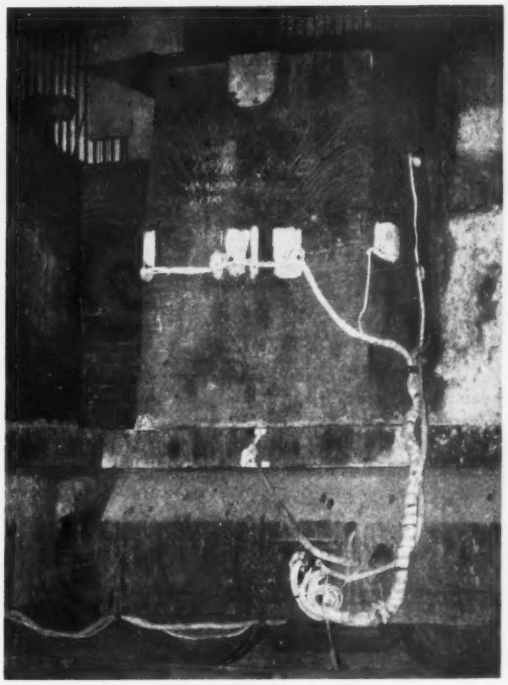


Fig. 2—Ingot mould, showing connections to thermocouple plugs

and added to the heat gain by the mould to give the heat loss from the steel.

Another method,[3] involving calculations from isotherms in the mould wall, gave similar results but was more cumbersome to apply. A typical set of isotherms is shown in Fig. 6. The results are tabulated in Table II and illustrated in Fig. 7.

Influence of Independent Variables—Consideration of the results for the 12 casts studied showed that the heat loss from the steel (H_s) increased roughly as the square root of the time from start of teeming (t, min.).

The variation of $H_s/\sqrt{t}$ with time was studied at eight times for each cast and it was found to fit closely to a quadratic equation, whilst its variation with initial mould temperature (θ, °C.) could be fitted only roughly to a linear regression.

The best-fit regression equation whose terms are all significant was found to be:

$$\frac{H_s}{\sqrt{t}} = 6{\cdot}88 - 0{\cdot}354\,\frac{\theta}{100} + 0{\cdot}482\left(\frac{t}{10}\right) - 0{\cdot}0304\left(\frac{t}{10}\right)^2$$

which fits the data with a standard deviation of 0·22.

An approximation to this is

$$\frac{H_s}{\sqrt{t}} = 7{\cdot}7 \qquad \sigma = 0{\cdot}59 \ (H_s \text{ in cal./g.})$$
$$= 0{\cdot}30 \qquad \sigma = 0{\cdot}021 \ (H_s \text{ in therms/ton}).$$

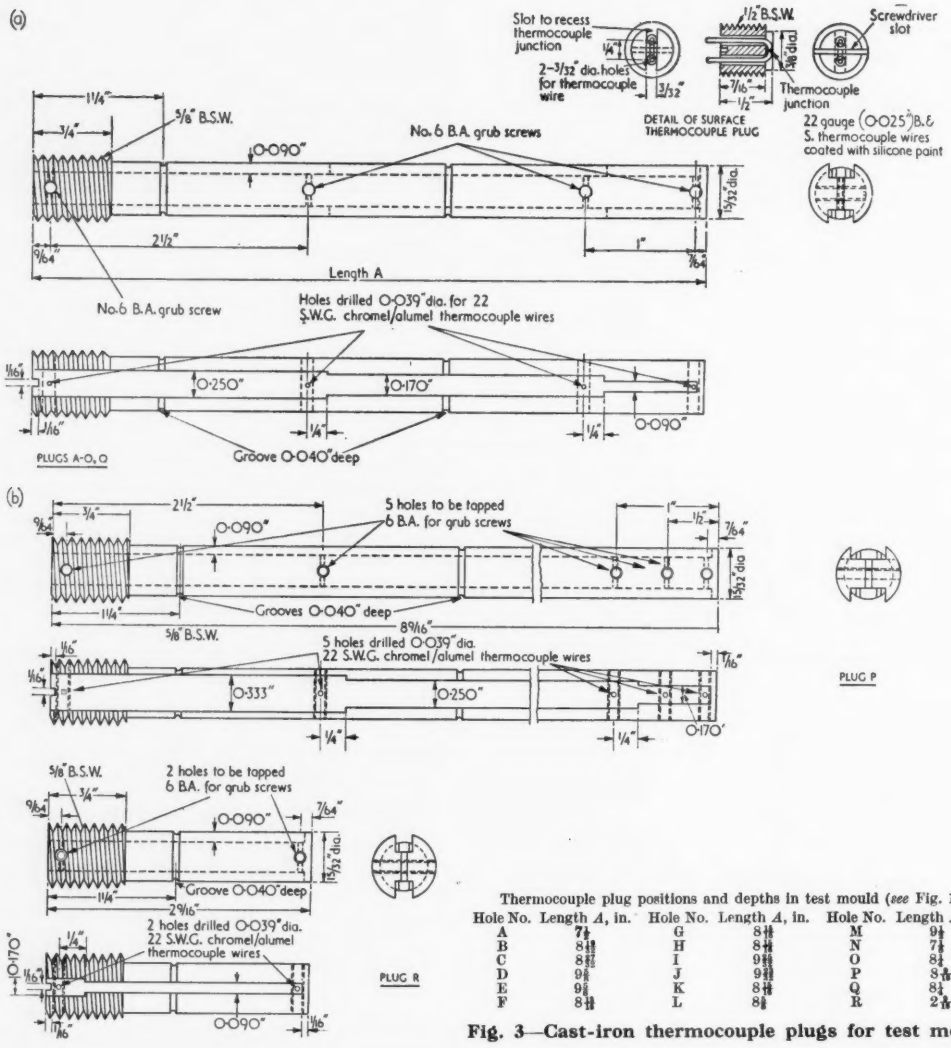
As was found with the 8-ton ingots previously reported, the effect of increasing initial mould temperature is to reduce the rate of heat loss. This effect is apparent in Fig. 8, but a further series of trials will be required to obtain a better estimate of its magnitude.

It is to be expected that the rate of cooling of ingots decreases with increase of ingot size, owing to

Table I
SUMMARY OF CONDITIONS FOR TEST MOULD

Trial No.	Cast No.	Date of Tap	Time of Tap	Time from Tap to Teem,* min.	Time from Teem to Strip, min.	Time from Previous Tap to Tap hr. min.	Time from Previous Strip to Tap hr. min.	Initial Mould Temperature, °C.	Steel Temperature, °C.		Position of Mould on Train (Fig. 3)
									Furnace	Start of Teem (Estimated)	
1	G 69	May, 1953 6	11.40	19	136	...	...	49			1
2	L 815	6	22.18	20	177	10 38	8 3	206	1585	1550	1 reversed
3	G 71	7	12.58	23	147	14 40	11 28	170	1590	1553	2
4	L 817	8	03.04	20	124	14 6	11 16	184	1590	1555	2
5	N 478	8	18.12	19	124	15 8	12 44	148			3
6	J 45	9	07.42	22	191	13 30	11 7	186	1560	1524	3 reversed
7	K 948	9	23.40	25	170	16 0	12 25	155	1605	1566	5
8	M 491	10	12.30	28	152	12 50	9 35	211			5
9	L 822	11	04.10	34	129	15 40	12 45	167	1600	1555	10
10	L 823	11	17.00	31	99	12 50	10 7	196	1590	1547	10 reversed
11	K 953	12	04.18	34	118	11 15	9 5	215	1595	1550	11
12	H 959	13	13.05	35	105	32 47	30 15	49			11

* First splash in ladle to first splash in test mould


(a)
Slot to recess thermocouple junction
2–3/32" dia. holes for thermocouple wire
DETAIL OF SURFACE THERMOCOUPLE PLUG
Screwdriver slot
Thermocouple junction
22 gauge (0·025") B.& S. thermocouple wires coated with silicone paint
No. 6 B.A. grub screws
Length A
No. 6 B.A. grub screw
Holes drilled 0·039" dia. for 22 S.W.G. chromel/alumel thermocouple wires
Groove 0·040" deep
PLUGS A–O, Q
(b)
5 holes to be tapped 6 B.A. for grub screws
Grooves 0·040" deep
5 holes drilled 0·039" dia. 22 S.W.G. chromel/alumel thermocouple wires
PLUG P
2 holes to be tapped 6 B.A. for grub screws
Groove 0·040" deep
2 holes drilled 0·039" dia. 22 S.W.G. chromel/alumel thermocouple wires
PLUG R

Thermocouple plug positions and depths in test mould (*see* Fig. 1)

Hole No.	Length *A*, in.	Hole No.	Length *A*, in.	Hole No.	Length *A*, in.
A	7 1/2	G	8 11/16	M	9 1/2
B	8 1/2	H	8 11/16	N	7 1/2
C	8 27/32	I	9 5/16	O	8 1/2
D	9 1/8	J	9 11/16	P	8 1/16
E	9 5/8	K	8 11/16	Q	8 1/2
F	8 11/16	L	8 1/2	R	2 1/16

Fig. 3—Cast-iron thermocouple plugs for test mould

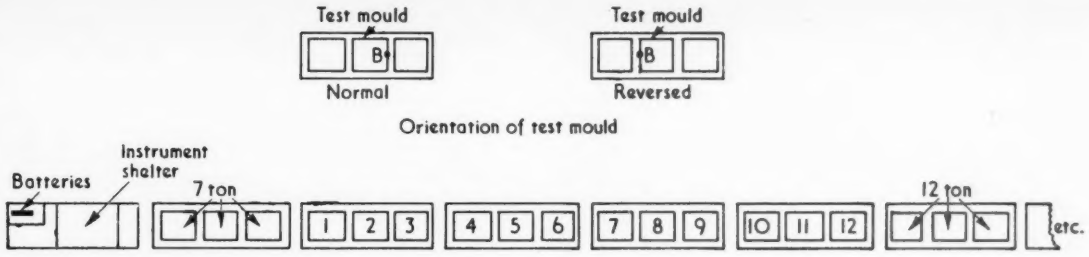


Fig. 4—Diagram of train, showing positions of 15-ton ingot moulds

an increase in the ratio of mass to surface area. This is confirmed by the data obtained from 8- and 15-ton ingots, as follows:

Ingot weight, tons	8	15
Mass of 1-cm. thick mid-height section, g.	48,300	76,160
Perimeter of mid-height section, cm.	350	432
Mass per unit area, g./sq. cm.	138	176
Mass per unit area ratio	1	1·28
$k(=H_s/\sqrt{t})$	9·76	7·70
Ratio	1·27	1

Rate of Solidification of Steel—Using the approximate relationship derived in a previous paper[3] for the distance from the steel surface solidified (S) for a given total heat loss, the rate of solidification of the steel at the mid-height section was calculated

Table II

TEST RESULTS, cal./g.

	Trial No.	Time from Start of Teem, min.										Initial Mould Temperature, °C.	$k(=H_s\sqrt{t})$
		5	10	20	30	40	50	60	80	100	120		
Heat loss from steel	1	15·0	21·7	32·2	41·8	49·7	57·2	63·5	74·5	83·8	91·1	49	7·79
Heat gain by mould		15·0	21·7	32·0	41·2	48·5	55·2	60·5	68·9	75·3	79·3		
Convection and radiation loss		0·0	0·0	0·2	0·6	1·2	2·0	3·0	5·5	8·5	11·8		
Heat loss from steel	2	16·1	21·7	31·7	39·4	46·1	53·4	59·5	68·1	75·8	82·0	206	7·35
Heat gain by mould		16·0	21·4	30·9	37·8	43·4	49·4	54·1	59·5	63·7	66·1		
Convection and radiation loss		0·1	0·3	0·8	1·7	2·8	4·0	5·4	8·6	12·1	15·9		
Heat loss from steel	3	17·4	21·7	30·5	39·2	47·0	55·2	63·5	73·3	81·2	88·2	170	7·55
Heat gain by mould		17·3	21·5	29·8	37·8	44·7	51·7	57·6	64·2	68·6	71·8		
Convection and radiation loss		0·1	0·2	0·7	1·4	2·3	3·5	5·9	9·1	12·6	16·4		
Heat loss from steel	4	14·8	20·4	31·6	40·4	47·9	55·8	62·2	71·9	80·0		184	7·51
Heat gain by mould		14·7	20·2	30·8	38·9	45·4	52·1	57·2	63·7	68·3			
Convection and radiation loss		0·1	0·2	0·8	1·5	2·5	2·6	5·0	8·2	11·7			
Heat loss from steel	5	17·4	23·1	34·0	43·3	51·3	59·6	66·2	75·2	83·3		148	8·06
Heat gain by mould		17·4	22·9	33·3	42·0	49·1	56·1	61·4	67·1	71·7			
Convection and radiation loss		0·0	0·2	0·7	1·3	2·3	3·5	4·9	8·1	11·6			
Heat loss from steel	6	13·3	19·4	29·0	37·8	46·4	54·2	61·7	71·6	80·2	87·0	186	7·29
Heat gain by mould		13·2	19·2	28·3	36·4	43·9	50·5	56·5	63·1	67·9	70·6		
Convection and radiation loss		0·1	0·2	0·7	1·4	2·5	3·7	5·1	8·5	12·3	16·4		
Heat loss from steel	7	16·2	21·1	31·9	41·6	49·7	58·0	65·3	76·5	84·6	93·0	155	7·83
Heat gain by mould		16·1	20·8	31·3	40·3	47·4	54·5	60·3	68·3	72·8	77·2		
Convection and radiation loss		0·1	0·3	0·6	1·3	2·4	3·5	5·0	8·2	11·8	15·8		
Heat loss from steel	8	15·5		31·7	39·7	49·1	56·6	62·4	72·1	79·8	86·5	211	7·58
Heat gain by mould		15·4		30·9	38·0	46·3	52·5	56·8	63·0	66·9	69·3		
Convection and radiation loss		0·1		0·8	1·6	2·7	4·1	5·6	9·1	12·9	17·2		
Heat loss from steel	9	14·2	20·1	31·0	41·0	49·4	57·7	64·6	74·1	81·1	87·2	167	7·65
Heat gain by mould		14·1	19·9	30·3	39·6	47·0	54·0	59·4	65·6	69·0	71·1		
Convection and radiation loss		0·1	0·2	0·7	1·4	2·4	3·7	5·2	8·5	12·1	16·1		
Heat loss from steel	10		20·9	30·6	39·6	46·9	53·9	61·6	71·4	79·0		196	7·35
Heat gain by mould			20·7	29·9	38·1	44·3	50·0	56·1	62·4	66·1			
Convection and radiation loss			0·2	0·7	1·5	2·6	3·9	5·4	9·0	12·9			
Heat loss from steel	11		21·6	32·1	41·3	50·0	57·3	64·0	72·8	79·9		215	7·67
Heat gain by mould			21·3	31·3	39·6	47·2	53·2	58·4	63·6	66·9			
Convection and radiation loss			0·3	0·8	1·7	2·8	4·1	5·6	9·2	13·0			
Heat loss from steel	12	18·4	24·7	35·1	44·5	53·4	60·6	67·1	75·5	(85·0)		49	8·28
Heat gain by mould		18·4	24·7	34·8	43·6	51·8	57·9	63·3	69·0				
Convection and radiation loss		0·0	0·0	0·3	0·9	1·6	2·6	3·8	6·5				
Mean heat loss from steel (H)		15·8	21·5	31·7	40·8	48·9	56·6	63·5	73·1	81·1	87·9		7·65
Standard deviation (σ)		1·6	1·5	1·6	1·9	2·2	2·2	2·2	2·3	2·7	3·5		
σ/H, %		10·1	6·8	5·0	4·6	4·4	3·9	3·4	3·1	3·3	4·0		

Fig. 5—Mould train, showing position of connecting cable

from the mean values of heat loss from steel given in Table II. The distance solidified can be expressed by the formula:

$$S = 2{\cdot}51\sqrt{t}\ \text{cm.}$$
$$= 0{\cdot}99\sqrt{t}\ \text{in.}$$

The intermediate and final results of this calculation are given in Table III and illustrated in Figs. 9 and 10.

Correlation with Mould Surface Temperature, and Heat Gained by a Single Plug—The heat gained by the mid-height section was plotted against the heat gained by plug *B* (Fig. 11*a*) and the temperature rise of a single surface thermocouple (Fig. 11*b*).

A good correlation was obtained with the former, which gave a non-linear relationship having a scatter of 3% about the mean value, but the latter gave a much greater spread of 10%.

Estimation of Overall Heat Balance for Mould and Bottom Plate

Only a very approximate estimate of the overall heat balance for the mould and bottom plate was

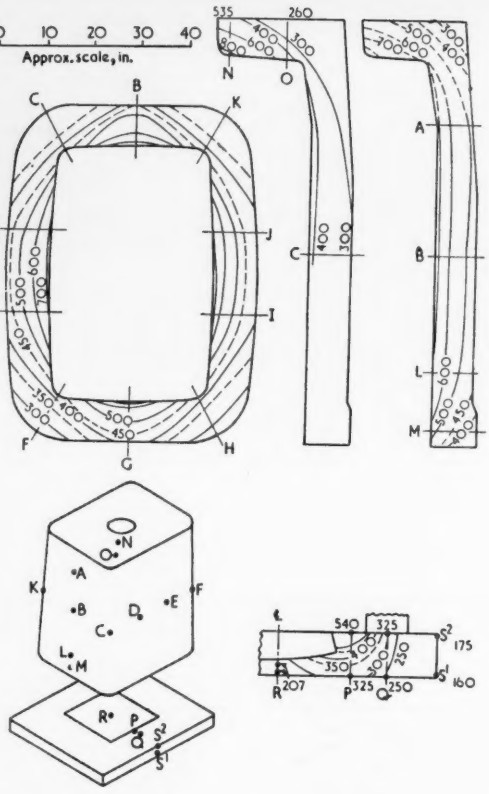


Fig. 6—Isotherms in wall of ingot mould in bottom plate: trial 1 (60 min.)

Table III

METHOD OF CALCULATION OF THICKNESS OF SOLIDIFIED STEEL (MEAN OF TRIALS 1–12)

Time from Start of Teem, min.	5	10	20	30	40	60	80	100	120
Heat loss from steel, cal./g.	15·8	21·5	31·7	40·8	48·9	63·5	73·1	81·1	87·9
Average inside surface temperature of mould θ_m, °C.	515	542	583	612	637	675	701	717	721
Instantaneous heat flux, cal./sq. cm. sec. $F_N = \frac{M}{60P} \cdot \frac{k}{2\sqrt{t}} = \frac{76,160}{60\times 432} \times \frac{7{\cdot}65}{2\sqrt{t}} = \frac{11{\cdot}24}{\sqrt{t}}$	5·03	3·56	2·51	2·05	1·78	1·45	1·26	1·12	1·03
Outward flux from steel, cal./sq. cm. sec. $\sigma(\theta_0+273)^4 = \frac{F_N}{\epsilon} + \sigma(\theta_m+273)^4$	10·58	7·72	5·75	4·93	4·59	4·00	3·74	3·55	3·39
Average steel surface temperature θ_0 °C.	1397	1271	1161	1107	1083	1036	1014	998	984
Temperature difference in steel $(1530-\theta_0)$, °C.	133	259	369	423	447	494	516	532	546
Coefficient of S^2 $L+C\frac{(\theta_1-\theta_0)}{6} = 65+0{\cdot}0267(1530-\theta_0)$	68·55	71·92	74·85	76·29	76·93	78·19	78·78	79·20	79·58
Coefficient of S $(a+b)[L+C\frac{(\theta_1-\theta_0)}{3}] = 7020+5{\cdot}76(1530-\theta_0)$	7786	8512	9145	9456	9595	9865	9992	10,084	10,165
Coefficient of S^0 $\frac{Q}{4\rho} = M\frac{(H_s-H')}{4\rho} = 2{\cdot}88\times 10^3(H_s-3{\cdot}2)$	36,288	52,704	82,080	108,290	131,616	173,664	201,312	224,350	243,936
Thickness of shell *S*, cm.	4·9	6·6	9·8	13·7	15·7	21·2	25·1	28·7	32·0
in.	1·9	2·6	3·9	5·4	6·2	8·3	9·9	11·3	12·6
% Solidified	17·9	23·7	34·2	45·9	51·4	65·3	73·9	80·6	86·3

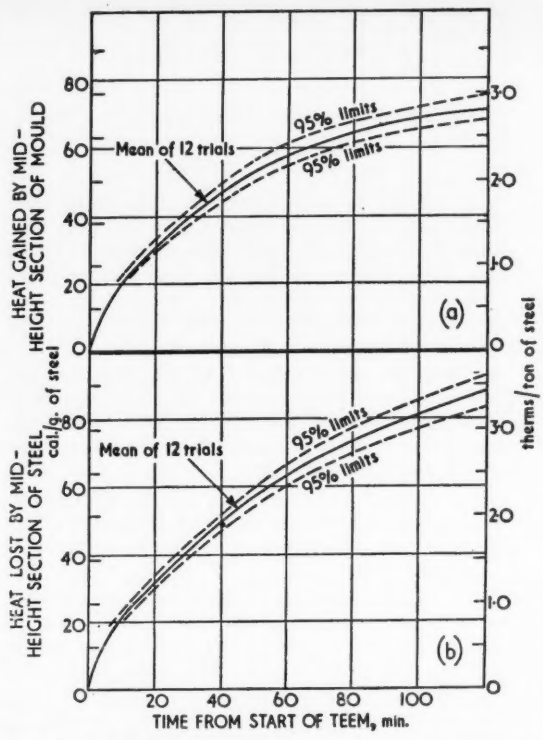


Fig. 7—Heat gained and lost *vs.* time: heat (*a*) gained and (*b*) lost by horizontal mid-height section of (*a*) mould, (*b*) steel

made, as there were insufficient thermocouples distributed in the mould to justify greater accuracy. The method used is described in Appendix II and the results are illustrated in Fig. 12. They suggest that for unit mass the average heat loss from the ingot as a whole is about 25% greater than the loss of the mid-height section.

A more detailed estimate could be made on a quarter section of the mould and bottom plate by increasing the number of thermocouples and plugs used at several horizontal and vertical sections. About 150 thermocouples would be required.

Rate of Cooling of Mould

As no method has yet been found of ascertaining the initial mould temperature with sufficient accuracy, except by using fixed thermocouples, an empirical rule has been deduced from measurements made on the cooling mould. Temperatures in the mould, though widely scattered at the time of stripping, were found to have evened out after 3 hr. Mould temperature θ_m could then be related to an exponential function of the time t' in hours from stripping to teeming:

$$\theta_m = 434 \cdot 5e^{-0 \cdot 0744t'}{}^{\circ}\mathrm{C}.$$

This relation is shown in Fig. 13.

APPLICATION TO WORKS PRACTICE

Owing to the limited amount of data and their inherent variability, it is not possible to make definite recommendations to modify existing works practice.

Nevertheless, it is known that stripping can be carried out safely at 100 min. from start of teeming and, in view of the marked flattening of the curve of heat loss after 60 min., it appears likely that ingots could be stripped after 80 min. This could be tested on a few casts, provided that the standing time can be reduced so that the moulds arrive earlier at the stripping bay.

On the average, the operating schedule and the conditions at the mid-height section of the first and last ingots of a cast, teemed and stripped in the same order, would approximate to Table IV.

The minimum standing time of the last ingot (35

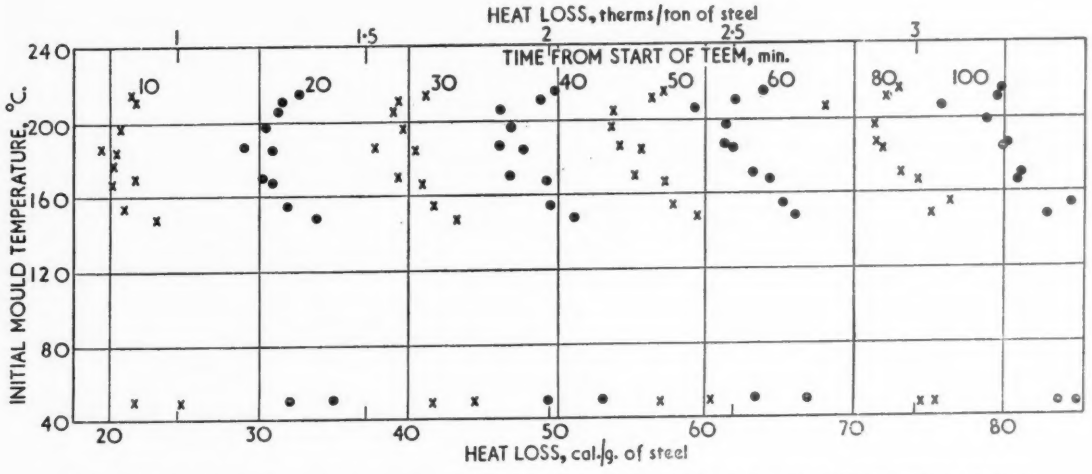


Fig. 8—Effect of initial mould temperature on heat loss from horizontal mid-height section of steel

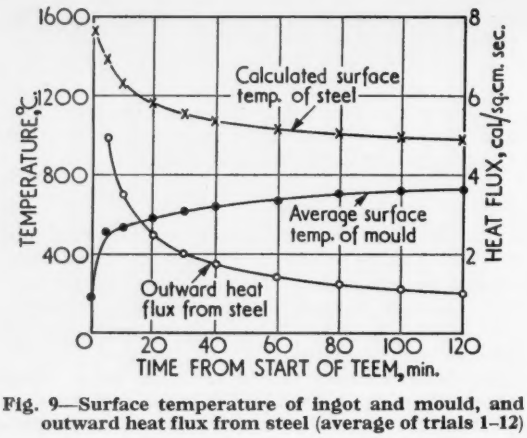


Fig. 9—Surface temperature of ingot and mould, and outward heat flux from steel (average of trials 1–12)

min.) is that quoted in the Productivity Report for three different sizes of rimming steel ingot. In making these suggestions, the effect upon ingot structure has not been taken into account.

CONCLUSIONS

(1) Using the mid-height section of the mould as criterion (*i.e.*, neglecting heat transfer perpendicular to that plane):

(*a*) The heat lost (H_s) from 15-ton bogie-cast bottle-top ingots after teeming can be expressed approximately by

$$\frac{H_s}{\sqrt{t}} = 7{\cdot}7, \quad \sigma = 0{\cdot}59 \ (H_s \text{ in cal./g.})$$
$$= 0{\cdot}30, \quad \sigma = 0{\cdot}021 \ (H_s \text{ in therms/ton}).$$

(*b*) The value of $H_s/\sqrt{t}$ varies with time, decreases with increase in initial mould temperature (θ° C.), and can be closely expressed by the regression equation

$$\frac{H_s}{\sqrt{t}} = 6{\cdot}88 - 0{\cdot}354\,\frac{\theta}{100} + 0{\cdot}482\left(\frac{t}{10}\right) - 0{\cdot}0304\left(\frac{t}{10}\right)^2, \quad \sigma = 0{\cdot}22.$$

(*c*) Making certain simplifying assumptions, the average thickness of solidified shell is given by the formula:

$$S = 2{\cdot}51\sqrt{t} \text{ cm.} = 0{\cdot}99\sqrt{t} \text{ in.}$$

(*d*) The temperature of the cooling mould after stripping can be estimated, after 3 hr., by the formula

$$\theta_m = 434{\cdot}5e^{-0{\cdot}0744t'} \text{ ° C.}$$

(*e*) The change in heat content of the whole mid-height slice can be estimated within 3% by a non-linear relationship, from the change in heat content of a single thermocouple plug. The corresponding relation with a single surface temperature is less satisfactory and gives an estimate of the heat content of the section accurate to 10%.

(2) From the limited data obtained in other parts of the mould and in the bottom plate it is estimated that the average heat loss per unit mass of the ingot as a whole is about 25% greater than the loss per unit mass of mid-height section.

(3) A further series of trials will be required to obtain a better estimate of the dependence of the heat loss from the mid-height section upon initial mould temperature.

(4) A reliable estimate of the total heat lost from the ingot can be obtained by temperature measurements on a quarter section of the mould and bottom plate, but 150 thermocouples will be required.

(5) The minimum stripping time for ingots can be reduced to 80 min. from start of teeming. This will necessitate a reduction in standing time.

Acknowledgments

The authors wish to express their thanks to the Steel Company of Wales, Ltd., for permission to undertake the work, and to the management, staff, and operators of the melting shop, the casting bay, the stripper bay, the mould preparation bay, the engineering department, and the metallurgical department for their assistance in preparing the test mould and carrying out the trials. They also wish to thank Mr. R. D. Collins and Mr. R. F. Slade, who undertook a large part of the statistical calculations, and Mr.

Table IV

OPERATING SCHEDULE AND CONDITIONS AT MID-HEIGHT SECTION

	Time, min.	Heat Lost by Steel, therms/ton		Minimum Thickness, in.	
		First Ingot	Last Ingot	First Ingot	Last Ingot
Tap	0	...	...	...	...
Start of teeming	20	0·2	...	0	...
Finish teeming	50	1·8	0·3	3·7	0
Leave casting bay	85	2·8	2·0	6·3	4·2
Arrive stripper bay	100	3·1	2·6	7·2	5·7
Finish stripping	115	(4·8)*	2·9	(11·2)*	6·6

* Stands stripped for 15 min.

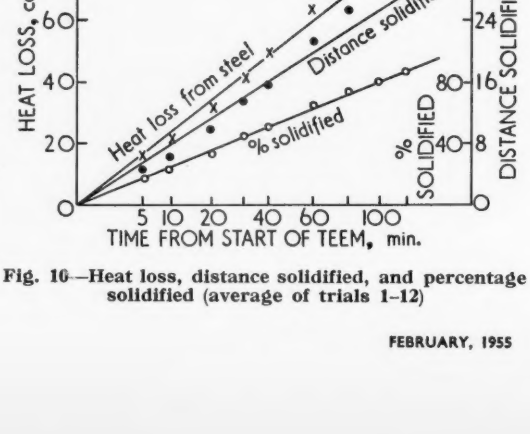


Fig. 10—Heat loss, distance solidified, and percentage solidified (average of trials 1–12)

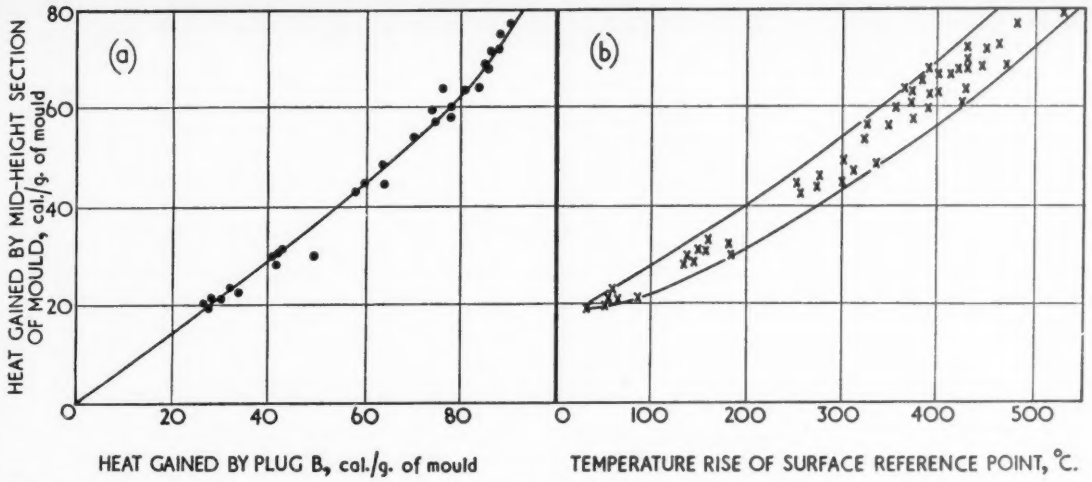

(a)
(b)
HEAT GAINED BY MID-HEIGHT SECTION OF MOULD, cal./g. of mould
HEAT GAINED BY PLUG B, cal./g. of mould
TEMPERATURE RISE OF SURFACE REFERENCE POINT, °C.

Fig. 11—Heat gained by mid-height section of mould *vs.* (*a*) heat gained by plug *B*, (*b*) temperature rise of surface reference point

R. S. Matfield and Mr. T. P. Conway for their assistance in making up the equipment.

References

1. L. H. W. Savage and R. T. Fowler: *J. Iron Steel Inst.*, 1953, vol. 173, pp. 119–128.
2. R. T. Fowler: *Instrument Practice*, 1952, vol. 6, pp. 223–241.
3. R. T. Fowler and L. H. W. Savage: *J. Iron Steel Inst.*, 1952, vol. 171, pp. 277–288.

APPENDIX I

Instrumentation of the Trials

The experimental programme called for the continuous recording of temperatures in the test mould and bottom plate and for intermittent recording of the temperature differentials (gradients) at their outer surfaces. The instruments used are shown schematically in Fig. 14 and illustrated in Fig. 15.

Mould temperatures were recorded by means of the 48-point switchbox (*A*) and 12-point high-speed electronic potentiometer (I) used in the earlier work.[3] Points 1, 14, 27, and 40 of the switchbox were used to measure the temperature at the junction boxes connecting the chromel/alumel thermocouples to the copper/constantan cable leading to the instrument cabin. As these junction-box temperatures differed only slightly from the ambient temperature, the numbers 1, 2, 3, and 4 were printed near the scale zero in successive cycles of the recorder, and served as ' markers ' to identify each set of 12 points.

Only 44 of the 60 mould thermocouples could be recorded in any run, in pairs, and these were selected by means of a plug panel, similar to a G.P.O. telephone switchboard, containing 24 4-pin sockets and 32 4-pin wander plugs. Each of these wander plugs was connected to two thermocouples, either the innermost or outermost pair of the four couples in each of the mould plugs shown in Fig. 3 or to a pair of ' markers.'

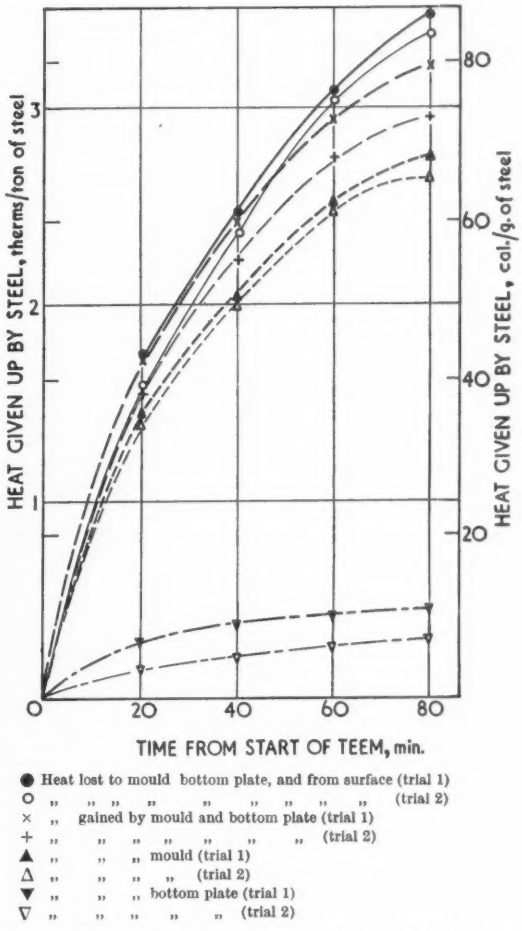

HEAT GIVEN UP BY STEEL, therms/ton of steel
HEAT GIVEN UP BY STEEL, cal./g. of steel
TIME FROM START OF TEEM, min.
● Heat lost to mould bottom plate, and from surface (trial 1)
○ „ „ „ „ „ „ „ „ „ (trial 2)
× „ gained by mould and bottom plate (trial 1)
+ „ „ „ „ „ „ „ (trial 2)
▲ „ „ „ mould (trial 1)
Δ „ „ „ „ (trial 2)
▼ „ „ „ bottom plate (trial 1)
∇ „ „ „ „ „ (trial 2)

Fig. 12—Estimated total heat lost from steel (trials 1 and 2)

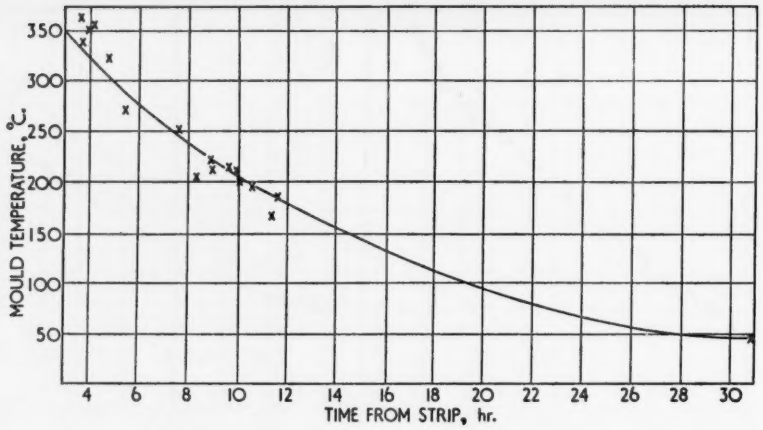


Fig. 13—Mould temperature at start of teem

The first 12 sockets were connected directly to the + and — points 1–24 of switchbox *A*, and were always used for inner pairs of couples and two markers. The second 12 sockets, used for outer pairs of couples and two markers, were connected to a multi-banked Oak-type switch (*C*), by which they could be connected either to points 25–48 of switchbox *A* for direct recording of temperature, or to switchbox *B* for recording differentials (*see* below). When recording differentials, a Post-Office springset, actuated by a cam on the switch *C*, closed an auxiliary circuit which 'homed' the uniselector in switchbox *A* after passing point 24.

The temperatures of the 12 thermocouples in the bottom plate were recorded by a 12-point medium-speed electronic potentiometer (II). The six innermost couples were connected directly to points 1–6 of this instrument, but the outer pair of couples were connected to a second multi-bank switch (*D*), by which they were connected either to points 7–12 for direct recording or to switchbox *B* for measurement of temperature differentials. When recording differentials, points 7–12 of recorder II were shorted out by means of a relay to prevent delay in printing when on open circuit.

The temperature differential between outer pairs of couples was frequently as low as 10° C. and, as the individual temperatures were recorded on a scale 0–1000° C. with an accuracy of about 2½° C., it was considered advisable to determine this differential more accurately on a more sensitive recorder. The thermocouple junctions were interconnected through the mould iron, and so the method used was to measure the potential first between the two chromel wires and then between the two alumel wires, using a single-point high-speed electronic recorder (III) with a centre zero and range — 2 to + 2 mV. By connecting the wires through the multi-bank switches *C* and *D* so that the deflections were alternatively + and —, the distance between each pair of deflections was equivalent to the temperature differential.

Selection of thermocouples for the recorder III was by means of switchbox *B*, similar in construction to switchbox *A*, but for which the impulse was provided by a micro-switch actuated by the chart drive mechanism of the recorder. The outer mould couples were connected *via* switch *C* to points 1–24 of box *B*, a second springset on this switch causing 'homing' of the uniselector after point 24 had been passed. The bottom plate couples were connected *via* switch

Table V

HEAT GAINED BY MOULD AND BOTTOM PLATE

Time from Start of Teem, min.		Mould Walls (therms)	Mould Top (therms)	Bottom Plate (therms)	Mould + Bottom Plate		Mould + Bottom Plate + Surface Loss (therms/ton of steel)
					Therms	Therms/ton of steel	
Trial 1	20	18·0	3·9	4·4	26·3	1·75	1·76
	40	25·5	5·3	5·5	36·3	2·42	2·48
	60	31·6	6·3	6·3	44·2	2·95	3·10
	80	34·4	6·9	6·9	48·2	3·21	3·47
Trial 2	20	17·7	3·4	2·3	23·4	1·56	1·60
	40	26·3	4·0	3·2	33·5	2·23	2·38
	60	32·3	5·0	4·0	41·3	2·76	3·04
	80	34·6	5·1	4·6	44·3	2·95	3·38

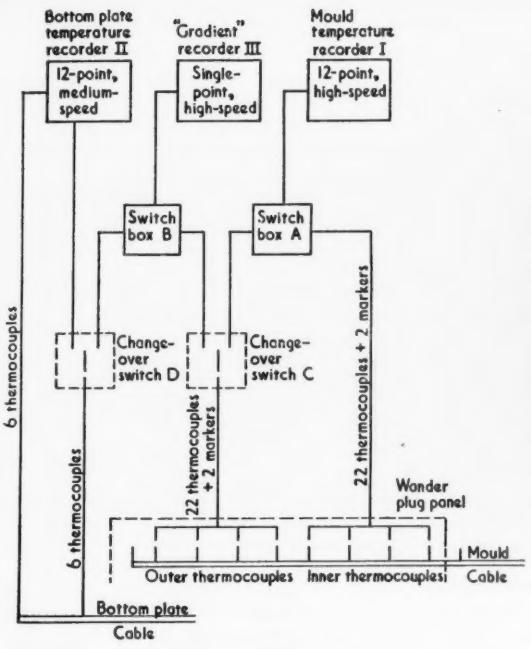


Fig. 14—Schematic arrangement of recording instruments and switchboxes

D to points 25–30 of box *B*, springsets on this switch maintaining operation of the switchbox only between these points.

Warning lamps were also incorporated in the control panel to ensure against switching both mould and bottom plate thermocouples on to differential recording at the same time.

The recorder III was also used for one of the spare inner-face thermocouples, particularly during the first 5 min. from start of teem, and for this purpose its range had to be changed, from — 2 to + 2 to 0–40 mV. This was accomplished by building the groups of resistors determining each range into blocks soldered to 8-pin plugs. The appropriate block could then be plugged into a socket fixed inside the recorder. The thermocouple connection was made through two spare pins of the 40-mV. resistor block and a further 4-pin socket mounted on the main panel.

Provision was made to supply power to the instruments and the switchboxes both when the train was stationary and whilst it was being shunted from one part of the works to another.

The recorders required 110 volts A.C. at 50 cycles and the switchboxes 24 volts D.C. When the train was stationary in the casting bay or mould preparation bay 110-V. A.C. mains were used directly for the recording instruments, and a supply was taken *via* a transformer-rectifier set for the switchboxes.

When the train was mobile a motor generator operated from a 60-V. battery supply was used to supply 110 volts A.C., 50 cycles, and another 24-V. battery supply served for the switchboxes.

The changeover from mains to batteries was controlled by a mains switch on the power control panel.

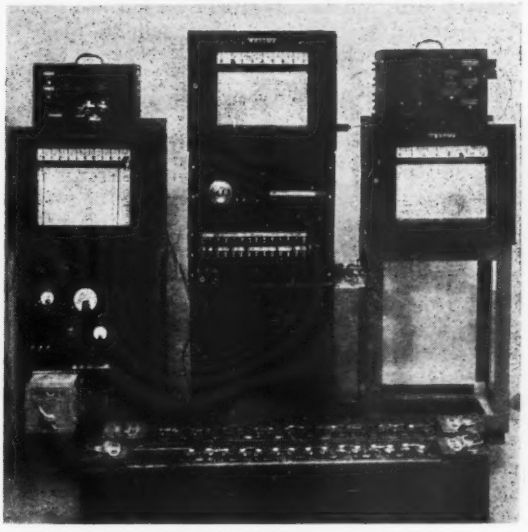


Fig. 15—Instrument assembly

To allow for any variation in recorder chart speed due to variations in mains or generator frequency, a second pen was used to mark the chart at 1-min.

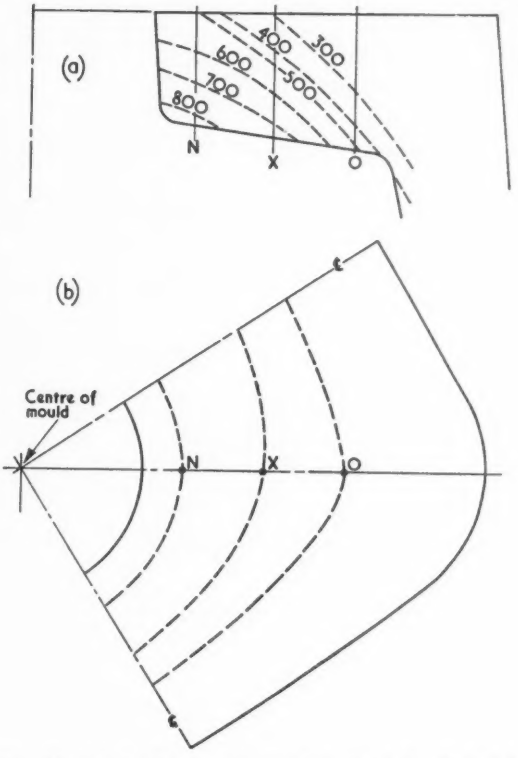


Fig. 16—(*a*) Isotherms in section through diagonal *NXO*; (*b*) probable positions of isotherms in mould top through *N*, *X*, and *O*

intervals. This pen was operated by a solenoid which was energized momentarily every minute when a clockwork timer closed a switch.

APPENDIX II

Estimate of Average Heat Loss from Whole Ingot

This estimate was made by considering the surface heat loss and the heat gained by three parts—mould walls, mould top, and bottom plate—the weight of each part being assessed from drawings.

Mould Walls

As a first approximation it was assumed that in any cross-section, the heat gained by a plug at a suitable reference point was in constant ratio to the heat gained by the whole section. Integrating this relationship over the total height of the mould walls, it is apparent that the heat gained by a vertical slice through the reference point was in the same ratio to the heat gained by the mould walls: *i.e.*, the ratio

$$\frac{\text{Heat gained by a 1-sq. cm. cross-section through plug } B}{\text{Heat gained by a 1-cm. thick mid-height section}}$$

was equal to the ratio

$$\frac{\text{Heat gained by a 1-cm. thick vertical section through } B}{\text{Heat gained by mould walls}}$$

The isotherms in the vertical section through B were plotted at intervals of 100° C. from observed temperatures in plugs L, M, B, and A (Fig. 6).

From the areas between the isotherms and the appropriate heat content of the mould iron, the total heat content of the vertical section and thus the heat gained by the mould walls was determined.

Mould Top

The isotherms in the mould top were plotted at intervals of 100° C. in the vertical section through N and O (Fig. 16*a*). This section was divided into 4 parts by the arbitrary insertion of a further line X and the heat gained by 1 sq. cm. thickness was determined from the areas between the isotherms and the appropriate heat content, as before.

In the plan of the top (Fig. 16*b*) the probable positions of three isotherms through N, X, and O were drawn. These divided the top, which was assumed to be a uniform 10 in. thick, into four zones corresponding to the four parts of section NXO, and from the area between the isotherms the weight of each zone was obtained.

The heat gained by the mould top was then determined as a first approximation from the weight of each zone and the heat gained (in cal./g.) by the appropriate part of section NXO.

Bottom Plate

A technique similar to that used for the mould top was used for the bottom plate, the vertical section in this case being $PQRS$ (Fig. 6).

Surface Heat Loss

To the total heat gained by the mould and bottom plate a small addition was made for the heat loss from the surface. This correction was assumed to be in the same proportion to the heat gained as that calculated for the mid-height section at the same time.

The results for trials 1 and 2 are tabulated in Table V and shown in Fig. 12.

The Formation and Decomposition of Hercynite ($FeO.Al_2O_3$)

By B. G. Baldwin, B.Sc., Ph.D., A.R.I.C., D.I.C.

SYNOPSIS

An investigation of the rate of formation of hercynite from mixtures of ferrous oxide and alumina at temperatures from 800° to 1300° C. and the effect of addition of lime, silica, magnesia, and dolomite on the formation of hercynite has been carried out. Most of the experiments were conducted on mixtures of 200–300 B.S.S. mesh ferrous oxide and α alumina in the correct proportions to yield hercynite. Experiments were also conducted with materials of different mesh size and on mixtures of ferrous oxide and γ alumina. 1011

DURING a previous investigation[1] into the formation and decomposition of fayalite, it was observed that alumina, when added to a mixture of ferrous oxide and silica, combined with the ferrous oxide when the mixture was heated at temperatures above 1050° C.

The presence of hercynite in the form of a solid solution with magnetite has been presumed in commercial sinters.[2] Alumina is a constituent of a large number of iron ores, and formation of hercynite during the reduction of ores in the blast-furnace is to be expected.

The formation of hercynite is of interest, as it is very much more difficult to reduce to metallic iron than ferrous oxide. Furthermore, the melting point is high (> 1750° C.) so that it will not take part in the slag type of bond in sintering which occurs, for example, with fayalite.

Lovell, Rigby, and Green[3] showed that when ferrous oxide was melted on alumina, distortion and cracking of the specimen occurred. This was due to the

Paper IM/B/18/54 of the Process Committee of the Ironmaking Division of the British Iron and Steel Research Association, received 17th June, 1954. The views expressed are the author's, and are not necessarily endorsed by the Committee as a body.

Dr. Baldwin is in the Blast Furnace Laboratory of the Imperial College of Science and Technology.

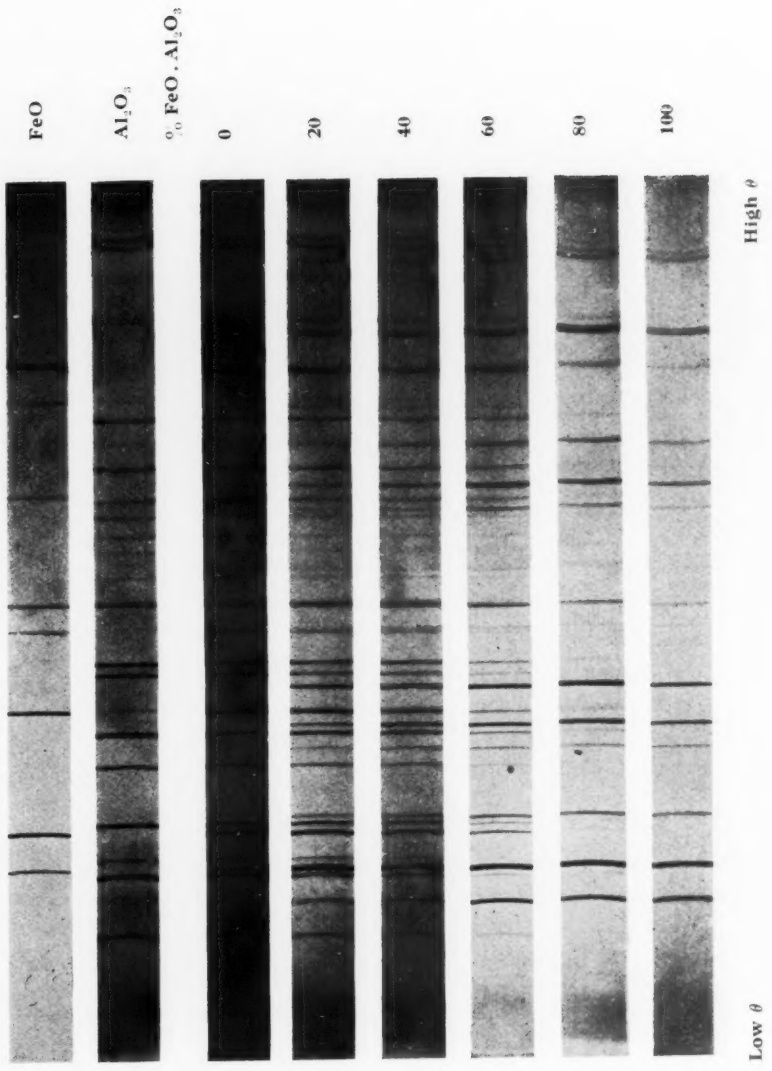


Fig. 1—X-ray powder photographs showing the change in pattern of a mixture of ferrous oxide and alumina with varying proportions of hercynite

expansion of the hercynite which was formed in the interior of the alumina specimen. Break-up of iron ores during reduction or during sintering may be expected to occur, since the formation of hercynite from alumina and ferrous oxide is accompanied by an increase in volume amounting to 57%.

EXPERIMENTAL TECHNIQUE

The experimental technique remained unchanged from that described in an earlier paper.[1] The mixtures were heated in iron crucibles in a nitrogen atmosphere for known times and at known temperatures, cooled rapidly, and analysed by an X-ray powder-diffraction technique. A series of X-ray powder photographs of mixtures of ferrous oxide and alumina, to which had been added known amounts of hercynite, was prepared, and, by visual comparison of the photograph of the treated mixture with this set of standard pictures, the amount of hercynite formed could be found to within 5%. Figure 1 shows the powder photographs of ferrous oxide and alumina and the change in pattern of a mixture of ferrous oxide and alumina with increasing amounts of hercynite. A 9-cm. camera and CoK_α radiation were used for all these photographs. An iron filter was used in all cases where the presence of weak β lines was undesirable. A number of microscopic examinations of the treated mixtures, which confirmed the conclusions drawn from the X-ray diffraction examination, were carried out. The sources from which the materials were obtained were described in the investigation into the formation of fayalite.

RESULTS

The effect of time of heating and temperature on the rate of formation of hercynite from a mixture of 200–300 mesh ferrous oxide and alumina is shown in Fig. 2*a*. No hercynite was detected after heating for 24 hr. at 700° C. The lengths of time when iron ore is at a high temperature during sintering and in the lower part of the blast-furnace are about $\frac{1}{4}$ and 2 hr., respectively, so that only the formation of hercynite in such short times will be of practical significance. One of the most interesting facts arising from these results is that the amount of hercynite formed at any given temperature increases with time. This was not found in the previous work done on the formation of fayalite; the amount of fayalite formed after heating a mixture of ferrous oxide and silica for $\frac{1}{4}$ hr. was approximately the same as that formed after heating for 24 hr. The reason for the difference in behaviour between the FeO–SiO_2 and FeO–Al_2O_3 systems is that in the former the rate of diffusion of ferrous oxide and silica through fayalite is very slow, and so once a layer of fayalite has formed between ferrous oxide and silica particles the reaction will practically cease. In the FeO–Al_2O_3 system, however, alumina and possibly ferrous oxide can diffuse at a finite rate through hercynite, and so the reaction is not stopped by a layer of hercynite forming between ferrous oxide and alumina particles. The amount of fayalite formed can be controlled by restricting the final temperature, regardless of the time of heating, whereas the amount of hercynite formed depends both on the time of heating and the temperature. The amount of fayalite formed when a mixture of 200–300 mesh ferrous oxide and silica is heated at different temperatures has been reported previously,[1] and the results are:

Temp., ° C.	900	1000	1100	1200
Fayalite, %	10	25	40	95

At 1200° C. the mixture had fused. Comparing these values and those obtained in the present investigation (shown in Fig. 2*a*), it will be seen that for times of up to 2 hr. the amount of fayalite formed is greater than that of hercynite. The low melting point of fayalite (1205° C.) compared with that of hercynite (> 1750° C.) will also favour the formation of fayalite at high temperatures. The reaction between ferrous oxide and silica might be expected to take place more readily during sintering and during reduction in the blast-furnace than the analogous ferrous-oxide–alumina reaction. However, it must be remembered that where ferrous oxide, silica, and alumina coexist, the formation of hercynite rather than fayalite occurs at temperatures above 1050° C.[1]

The formation of hercynite during the heating of ferrous chamosite has been studied by Brindley and Youell.[4] Hercynite was formed during heating in steam or vacuum (which prevents oxidation to the ferric form) at temperatures from 600° C. and above. This is of considerable interest in the study of the blast-furnace process, since ferrous chamosite is a constituent of several British ores. In the present investigation no hercynite was formed until the temperature was above 700° C., and this is no doubt due to the ferrous iron and alumina being in a far greater state of division in natural chamosite than in the artificial mixture used in this investigation.

Effect of Crystalline Form of Alumina

Alumina is often present in iron ores in hydrated form which on heating changes to γ alumina, and Fig. 2*b* shows the rate of formation of hercynite at 800°, 900°, 1000°, 1100°, 1200°, and 1300° C. from 200–300 mesh ferrous-oxide–γ-alumina mixtures. Comparison with Fig. 2*a* shows that at temperatures below 1200° C. the rate of formation of hercynite is more rapid when the alumina is in the γ form than in the α form. The rate of formation at 1200° and 1300° C. is apparently the same as for both forms of alumina. The powder-diffraction photographs showed that the alumina present in the final product was still in the γ form after heating at 900° C. for 24 hr. At 1000° C. γ alumina was present after 3 hr., but at 6½ hr. was converted to the α form. At 1100° C. the alumina was still present in the γ form after ½ hr. but after 3 hr. it was practically completely converted to the α form. At 1200° and 1300° C. the alumina was present as the α form after ½ hr., and the rate of formation of hercynite was the same as was found with mixtures of α alumina and ferrous oxide. At 1200° and 1300° C. the γ alumina is very rapidly changed to the α form, so that a mixture of γ alumina and ferrous oxide heated to these temperatures acts like one composed of α alumina and ferrous oxide; hence the rate of formation of hercynite is the same for alumina present as either the α or the γ form at the start of the heating. It may be seen that alumina in the γ form is more reactive

towards ferrous oxide than in the α form, and this may be of practical significance.

Effect of Particle Size

The effect of variation in the particle size on the rate of formation of hercynite with time is shown in Fig. 3 for experiments carried out with 200–300, 150–200, 120–150, and 100–120 B.S.S. mesh material (corresponding to mean particle sizes of 0·065, 0·090, 0·115, and 0·138 mm., respectively) at 1200° and 1300° C. The results obtained after heating for 3 hr. at 1000° C. are shown in the bottom right-hand corner. At 1200° and 1300° C. the amount of hercynite formed after a given time of heating increases with decreasing particle size. It is interesting to compare these results with those obtained in the investigation of the formation of fayalite. In the latter case little or no interaction took place when the particle size was larger than 200–300 mesh at temperatures up to the melting point (1205° C.). There would seem to be greater mobility in the $FeO–Al_2O_3$ system than in the $FeO–SiO_2$ system.

Effect of Addition of 10% of Magnesia

Mixtures of 200–300 mesh ferrous oxide and alumina to which had been added 10 wt.-% of magnesia of a similar particle size were heated for 3 hr. at 900°, 1000°, 1100°, and 1200° C. The results, together with those obtained in the absence of magnesia (free magnesia was present in all the samples) were:

Temp., ° C.	900	1000	1100	1200
Hercynite (normal mixture), %	10	30	50	90
Hercynite (mixture + 10% MgO), %	10	30	40	70

A slight shift of the lines due to hercynite on the powder photographs of the samples heated to 1200° C. was noticed. This was in such a direction as to indicate that the lattice parameter was decreasing,

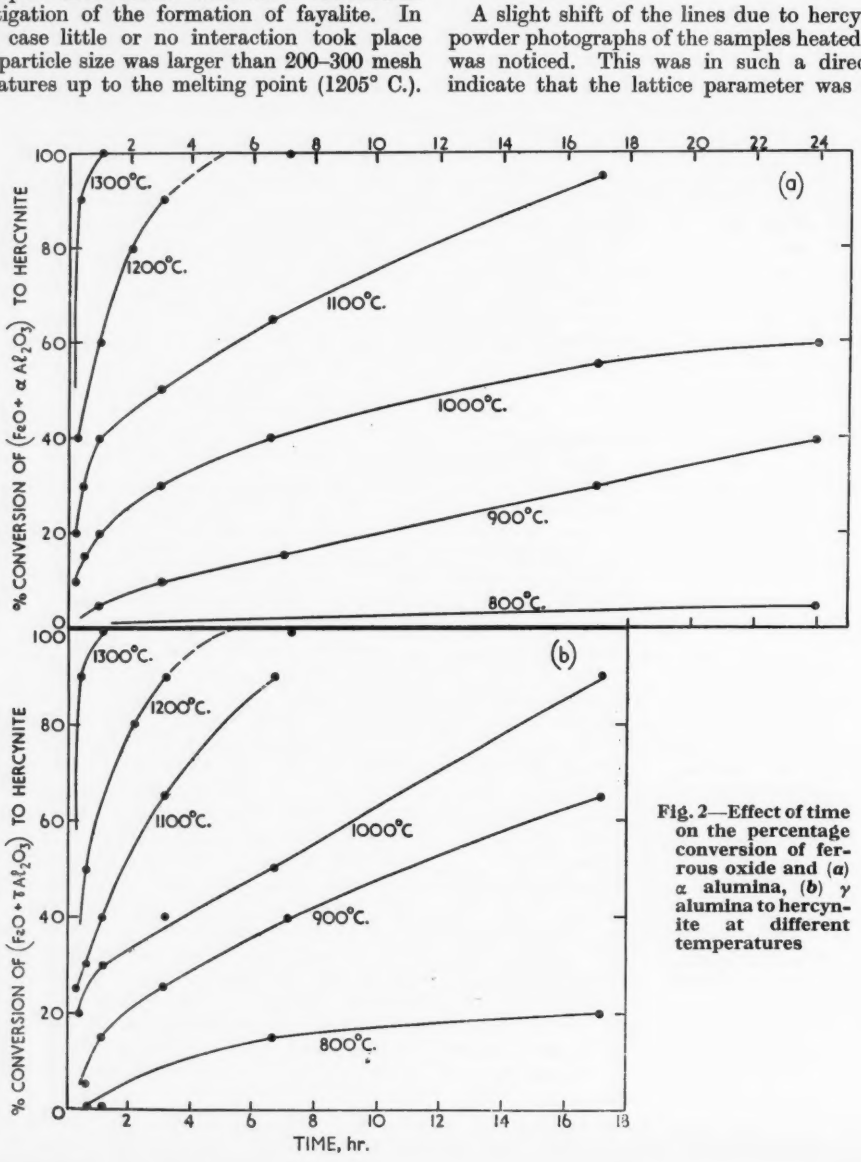


Fig. 2—Effect of time on the percentage conversion of ferrous oxide and (*a*) α alumina, (*b*) γ alumina to hercynite at different temperatures

owing to solid solution of the hercynite with an isomorphous substance of lower molecular weight. This is presumed to be due to the formation of a solid solution of hercynite and Spinel ($MgO.Al_2O_3$). The lattice parameter of the hercynite was changed from 8·143 to 8·123 (all the lattice parameters are given in kX. units). Assuming that Vegard's law holds for a mixture of Spinel and hercynite and using the value of 8·068 as the value of the lattice parameter of Spinel, this corresponds to a spinel of approximate composition 20% $MgO.Al_2O_3$, 80% $FeO.Al_2O_3$. The original amount of magnesia present was 10%, and this amount of spinel would be formed from just over half this amount. The amount of hercynite formed at 1200° C. and shown above has been corrected for solid solution of Spinel.

The experiment was repeated using 30% of magnesia instead of 10% and the mixture was heated to 1200° C. for 3 hr. as before. The powder photograph of the treated mixture showed a greater shift of line than before; the lattice parameter of the hercynite had been decreased to 8·113 and excess magnesia was present. This parameter corresponds to a spinel of approximate composition 35% $MgO.Al_2O_3$, 65% $FeO.Al_2O_3$. The amount of magnesia required for this is one-third of that available, and this accounts for the presence of excess magnesia in the final mixture. The formation of Spinel and hercynite is proceeding simultaneously at 1200° C. This is in agreement with the known thermodynamic data for the formation of hercynite and Spinel, since the free-energy changes are approximately the same.[5]

Effect of Addition of 10% of Lime

The experiments described were repeated using lime instead of magnesia. The time of heating was 3 hr. at each temperature. At 900° C. no hercynite could be detected by the X-ray method, and free ferrous oxide, alumina, and lime were present. At 1000° C. hercynite was formed and free ferrous oxide and alumina were present. All the lime had disappeared and a new crystalline phase, which was identified as a monocalcium aluminate, was present. Comparison of the strength of the ferrous oxide and hercynite lines showed that approximately 30% conversion to hercynite had taken place. The ratio of the intensity of the alumina lines to those of ferrous oxide was diminished, agreeing with the fact that lime was combined with some of the alumina. There was also very slight line shift of the hercynite pattern, signifying that the lattice parameter had been altered. At 1100° C. similar reactions had taken place, but there was considerably more hercynite present than at 1000° C., although less than was formed under similar conditions in the absence of lime. At 1200° C. hercynite, monocalcium aluminate, and free ferrous oxide were the only crystalline phases identified; the lattice parameter of the hercynite was found to be 8·160 (normal hercynite = 8·143).

Pure hercynite was next heated to 1200° C. with 10% lime. Examination of the heated material showed that it had changed colour from dark green to black. The X-ray diffraction photograph showed that there was hercynite present, although the lattice parameter had increased, and that monocalcium aluminate was present, but no lime could be detected. The lines due to ferrous oxide were diffuse, showing that it was in the form of very small crystals. The experiment was repeated, but this time the amount of lime was increased to give an Al_2O_3 : CaO molecular ratio of 1 : 1. On examining the heated mixture only a trace of hercynite was found, the remaining

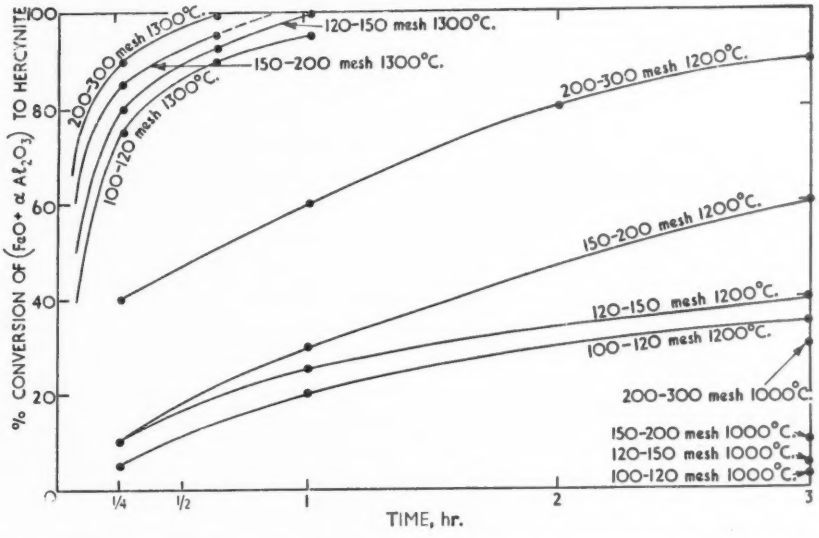


Fig. 3—Effect of variation in particle size on the percentage conversion of ferrous oxide and α alumina at 1000°, 1200°, and 1300° C.

crystalline constituents being ferrous oxide and monocalcium aluminate.

It is evident that alumina has a greater tendency to combine with lime than ferrous oxide, and that lime can completely displace ferrous oxide from hercynite.

The thermodynamic data for the formation of calcium aluminate are rather uncertain,[5] but the present results show that the free-energy change for the formation of monocalcium aluminate should be greater than that for the formation of hercynite. The change in the lattice parameter of the hercynite lattice from the normal of 8·143 to 8·160 is unexpected. It is unlikely to be due to solid solution with calcium aluminate, as this has the structure of a true oxygen salt whilst hercynite can be regarded as a mixed oxide, and hence solid solution between the two should be very slight, if at all. It is also unlikely to be due to magnetite, as it would require 9% of magnetite to cause this change, and this would require either the decomposition of the ferrous oxide into magnetite and iron (no iron was observed) or the oxidation of ferrous oxide, which, under the experimental conditions, was impossible.

Comparing these results with those obtained in the fayalite investigation, where the presence of 10% of lime prevented the formation of fayalite, it will be seen that the effect of lime is very much greater in the FeO–SiO_2 system than in the FeO–Al_2O_3 system. In both systems, however, sufficient lime will displace the ferrous oxide from combination with either the alumina or silica and so increase the ease of gaseous reduction of iron.

Effect of Addition of 10% of Ignited Dolomite

This was found to act as a mixture of lime and magnesia. The lime combined with alumina and the magnesia remained unchanged.

Effect of Addition of 10% of Silica

Mixtures of ferrous oxide and alumina to which was added 10% of silica (in the form of pure quartz) were heated at 900°, 1000°, 1100°, and 1200° C. for 3 hr. At 900° C. a small amount of hercynite was present (approx. 5%). The remaining crystalline material was composed of ferrous oxide, alumina, and silica. At 1000° C. the amount of hercynite had increased to about 20% and ferrous oxide and alumina were found, but the amount of silica was less than would be expected if none had reacted. At 1100° C. hercynite and alumina were present, no ferrous oxide could be detected, and only a trace of silica was found. At 1200° C. hercynite was the major crystalline constituent and a trace of alumina was the only other substance identified. The lattice parameter of the hercynite remained unchanged during all these experiments.

Pure hercynite was next heated with 10% of silica at 1200° C. The product was green in colour and consisted of normal hercynite (*i.e.*, lattice parameter = 8·143), together with a small quantity of an unidentified substance; no silica could be detected. The experiment was repeated with the amount of silica increased to give an $Al_2O_3 : SiO_2$ molecular ratio of 1 : 1. The product was green and consisted of normal hercynite and an aluminium silicate, either mullite or sillimanite (the quantities were so small that it was not possible to differentiate between these compounds). No crystalline ferrous oxide was identified, and it was assumed that it was probably present in a very fine state of subdivision. Comparison of this X-ray photograph with that obtained with a 10% addition of alumina showed that the unidentified substance was aluminium silicate.

The effect of silica on the formation of hercynite from the reaction between ferrous oxide and alumina is not absolutely clear. The formation of hercynite and aluminium silicate proceeds simultaneously, but the excess ferrous oxide does not always appear in the final product. When hercynite is heated with silica the formation of aluminium silicate occurs, and this is due to decomposition of the hercynite, but no ferrous oxide can be detected in the final product. This may be due to the ferrous oxide being present in a very fine state of subdivision (as was noticed with the experiments carried out with lime addition).

CONCLUSIONS

(1) Ferrous oxide and alumina combine at temperatures above 700° C. to give hercynite.

(2) The reaction of ferrous oxide with γ alumina is more rapid than that with α alumina over the temperature range where γ alumina is stable.

(3) Increase in the particle size of the ferrous oxide and alumina in the range 300–100 B.S.S. mesh decreases the rate of formation of hercynite.

(4) Magnesia, added to a mixture of ferrous oxide and alumina, acts primarily as an inert diluent towards the formation of hercynite. At 1200° C. Spinel ($MgO.Al_2O_3$) forms simultaneously and goes into solid solution with the hercynite.

(5) Lime, above 1100° C., restricts the formation of hercynite by combining with the alumina. Lime can completely displace ferrous oxide from hercynite at 1200° C.

(6) Dolomite acts as a mixture of lime and magnesia.

(7) The effect of addition of silica is uncertain, although it can decompose hercynite to give aluminium silicate. The reactions between alumina and silica and between alumina and ferrous oxide proceed simultaneously at 1200° C.

Acknowledgments

The author wishes to express his appreciation to Dr. H. L. Saunders, former head of the Ironmaking Division of the British Iron and Steel Research Association, for his helpful advice and interest in this work, and to Dr. R. Wild for assistance in the interpretation of the X-ray diffraction photographs.

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Sulphur Partition between Gas, Slag, and Metal Phases

UNDER IRON AND STEELMAKING CONDITIONS

By E. T. Turkdogan, Ph.D., M.Met.

SYNOPSIS

The sulphur reaction in iron and steelmaking processes has been examined on the basis of available gas–slag–metal equilibrium data. The sulphur capacity of the slag, *i.e.*, $\mathfrak{C}_S = (N_S)[a_O/a_S]$ is related to $\lambda = 1/(N_{SiO_2} + 1{\cdot}5N_{P_2O_5} + 1{\cdot}5N_{Al_2O_3})$ for a given temperature. The sulphide capacity C_S is also related to λ.

Both log $\mathfrak{C}_S$ and log C_S are linear functions of $1/T$ for a given value of λ. Under reducing conditions, *e.g.*, blast-furnace practice, the (S%)/[S%] ratio is related to the [Mn%]/(Mn%) ratio. The application of the present considerations to works' data indicates that in iron and steelmaking processes the sulphur transfer to and from the slag and metal phases reaches equilibrium. The sluggishness of the carbon–oxygen reaction, as compared with the others taking place at the slag/metal interface, has been discussed. 1048

Introduction

A NUMBER of equations have in the past been suggested for evaluating the equilibrium sulphur distribution between slag and metal during steelmaking. It was shown by Schenck[1] and McCance[2] that the sulphur partition, (S%)/[S%], could be related to the manganese content of the iron, and the ferrous oxide, calcium oxide, and silica contents of the slag; they also indicated that for a given (CaO%)/(FeO%) ratio and silica concentration, the sulphur partition increased with temperature.

Later experiments by Grant and Chipman[3] showed that the sulphur distribution factor could be related to the excess base or acid in the slag. They calculated the 'excess base or acid' term as follows: "The basic oxides, CaO, MgO, and MnO, were taken in all ranges to be equal on a mole per mole basis. The following molar ratios were taken as necessary for a neutral slag:

2 Base: $1SiO_2$; 4 Base: $1P_2O_5$; 2 Base: $1Al_2O_3$; 1 Base: $1Fe_2O_3$

When insufficient base was available, Fe_2O_3 was considered neutral and added to the FeO. FeO was considered uncombined in all cases. The remaining base or acids, after these combinations were fulfilled, were then added and designated as the 'excess base or acid count,' expressed in mols per 100 g. of slag." Although the method adopted by Grant and Chipman was empirical and concealed certain facts concerning the sulphur reaction, some interesting and useful information was brought to light. For example, calcium oxide, magnesium oxide, and manganous oxide are replaceable on a mole per mole basis, and fluorspar is neutral and does not affect the slag/metal sulphur partition.

Samarin, Temkin, and Schwarzmann[4] have shown that the overall reaction controlling the distribution of sulphur between iron and basic slags may be written as

$$[S] + (O) = (S) + [O] \quad \ldots\ldots(1)$$

where square and round brackets indicate reacting species dissolved in iron and slag, respectively. Herasymenko and Speight[5] have evaluated the equilibrium ratio of this reaction on the assumption that slags are fully ionized. Their method is, however, of restricted value since there is no way of calculating the oxygen-ion concentration except in highly basic slags.

Arguing on the basis of similarity, Richardson and Withers[6] assumed that the distribution of sulphur between gas and any molten slag would obey the relationship

$$\tfrac{1}{2}S_2(g) + (O) = (S) + \tfrac{1}{2}O_2(g) \quad \ldots\ldots(2)$$

and this assumption has now been verified experimentally by Richardson and Fincham[7, 8] for oxygen pressures below about 10^{-6} atm. At higher oxygen pressures, *e.g.*, above 10^{-4} atm., sulphur exists in the slag mainly as sulphate but Richardson and Fincham have shown that there is no need to take account of sulphate formation when discussing iron and steelmaking processes.

Existing experimental data are not suitable for calculating the equilibrium constants of reactions (1) and (2) since, as mentioned above, there is no generally valid method of determining the oxygen-ion content of a slag; however, these constants are able to be used in the following way. The equilibrium constant for reaction (1) is given by

$$K = \frac{(a_S)[a_O]}{[a_S](a_O)} \quad \ldots\ldots(3)$$

At the low concentrations of slag sulphur usually encountered its activity coefficient is not likely to vary with slag composition and for any slag (a_O) may be considered constant. Equation (3) may therefore be written as

$$\mathfrak{C}_S = (N_S)[a_O/a_S] \quad \ldots\ldots(4)$$

where (N_S) = ion fraction of sulphur in the slag

(a_O) = activity of oxygen in the metal $(= f_O[O\%])$

$[a_S]$ = activity of sulphur in the metal $(= f_S[S\%])$.

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Dr. Turkdogan is Head of the Physical Chemistry Section of the Chemistry Department of the Association.

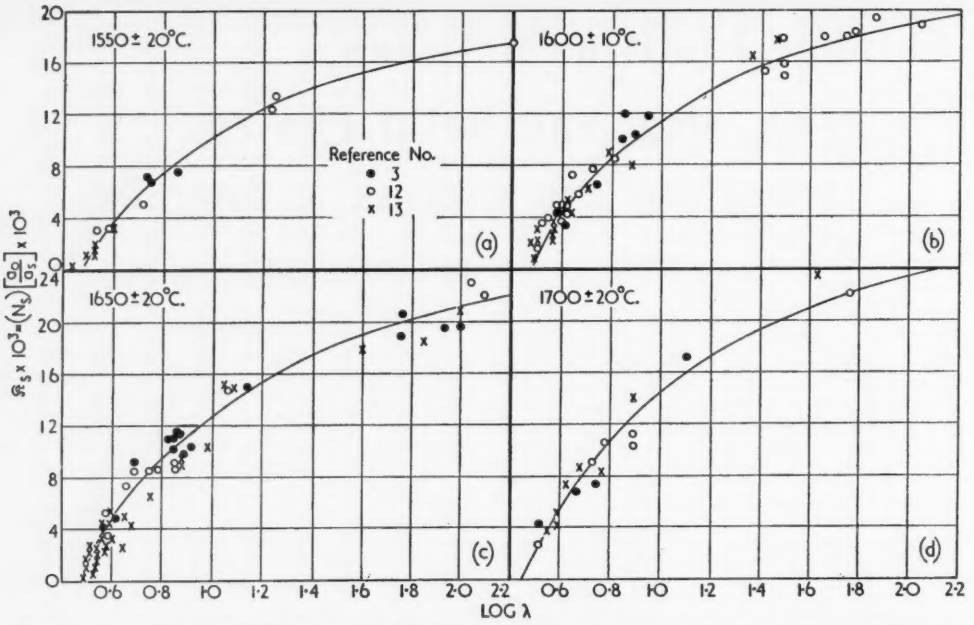

Fig. 1—Sulphur capacity related to slag composition at mean temperatures of (*a*) 1550 ± 20° C.; (*b*) 1600 ± 10° C.; (*c*) 1650 ± 20° C.; (*d*) 1700 ± 20° C.

The effects of dissolved elements on the activity coefficients f_O and f_S are known for a number of solutions.[9, 10] When the oxygen content of the metal is not known, the value of $[a_O]$ can be calculated from the slag composition by using the ferrous oxide isoactivity diagram put forward by Turkdogan and Pearson.[11]

By analogy with the expression suggested by Richardson and Fincham,[7, 8] and mentioned later, $\mathfrak{K}_S$ may be called the sulphur capacity of the slag. For any system in which the metal and slag have been equilibrated, the sulphur capacity may be evaluated from a knowledge of the compositions of the metal and slag.

Richardson and Fincham[7, 8] have treated equation (2) in a corresponding manner to obtain the sulphide capacity of the slag, *i.e.*,

$$C_S = (S\%)\sqrt{p_{O_2}/p_{S_2}} \quad \ldots\ldots(5)$$

It seems preferable to use the ion fraction of sulphide rather than the percentage concentration and in later discussions the sulphide capacity will be taken as

$$C_S = (N_S)\sqrt{p_{O_2}/p_{S_2}} \quad \ldots\ldots(6)$$

The terms sulphur capacity $\mathfrak{K}_S$ and sulphide capacity C_S have been used to denote that the former is calculated from the sulphur and oxygen contents of the metal and the latter from the sulphur and oxygen pressures in the gas phase. The two are of course interrelated.

EQUILIBRIUM SULPHUR PARTITION BETWEEN MOLTEN IRON AND SLAG

Steelmaking Conditions

The best experimental results for the study of the sulphur equilibrium under slags corresponding to steelmaking practice are those given by Grant and Chipman,[3] Fetters and Chipman,[12] and Winkler and Chipman.[13] Table I summarizes their data and indicates the slag and metal composition ranges studied. It will be seen from equations (3) and (4) that the sulphur capacity $\mathfrak{K}_S$ is a function of the oxygen-ion activity of the slag and since the latter will be influenced by the slag composition, $\mathfrak{K}_S$ should in turn also be a function of the slag composition. A careful examination of the experimental results of Chipman and co-workers[3, 12, 13] indicates that $\mathfrak{K}_S$ can be related to the 'acids' content of the slags. Because $\mathfrak{K}_S$ increases as acidity decreases, it is convenient to introduce the acidity as an inverse function. It has been found that $\mathfrak{K}_S$ is simply related to what

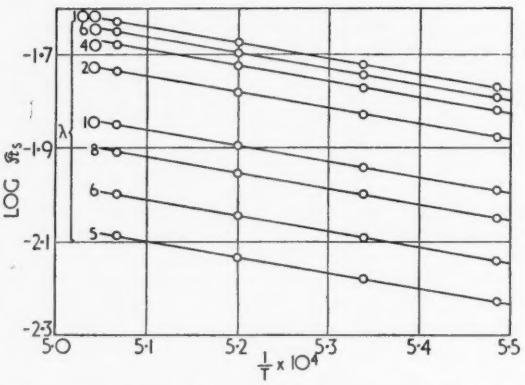

Fig. 2—Variation of logarithm of sulphur capacity with temperature

Table I

SLAG–METAL EQUILIBRIUM DATA

From the work of Chipman *et al.*[3,12,13]

Ref.	Temp., °C.	Metal Analysis, wt.-%				Slag Analysis, mole fraction									
		Mn	P	S	O	CaO	MgO	MnO	FeO	Fe_2O_3	SiO_2	P_2O_5	Al_2O_3	S	F
3	1540–1670	0·0	0·0	0·08–0·20	...	0·20–0·47	0·09–0·16	0·00	0·29–0·48	0·015–0·052	0·02–0·14	0·00	0·03–0·14	0·004–0·018	0·00
12	1540–1720	0·0	0·0	0·017–0·054	0·003–0·300	0·00–0·47	0·04–0·26	0·00	0·003–0·920	0·001–0·050	0·01–0·30	0·00	0·00	0·0005–0·0030	0·00
13	1540–1720	0·0–0·4	0·0–0·18	0·016–0·075	0·003–0·250	0·00–0·47	0·04–0·39	0·00–0·36	0·005–0·580	0·002–0·032	0·01–0·36	0·00–0·08	0·00–0·07	0·0002–0·0060	0·00–0·24

may for convenience be called the ' sulphur capacity index ' (λ) where

$$\lambda = 1/(N_{SiO_2} + 1{\cdot}5N_{P_2O_5} + 1{\cdot}5N_{Al_2O_3}) \quad \ldots\ldots\ldots(7)$$

The values of C_S derived from the data referred to above are plotted in Figs. 1*a–d* against the appropriate values of log λ for various mean temperatures. Within the range of compositions studied, all attempts to discern the influence of the relative proportions of the various basic oxides, CaO, MgO, etc., were unsuccessful. This confirms the statement of Grant and Chipman[3] that the basic oxides are equivalent on a mole per mole basis.

The results quoted by Grant and Chipman do not include the oxygen contents of their metal samples; [a_O] was therefore calculated from the slag composition[11]; the corresponding values of C_S agree well with those for which [O%] was determined. The results of Winkler and Chipman[13] show that fluoride ions do not affect the C_S/log λ relationship.

The effect of temperature on sulphur capacity is best illustrated by a plot of log C_S against $1/T$, some examples of which are shown in Fig. 2. The parallel lines may be represented by the equation:

$$\log C_S = -(3380/T) + \nu \quad \ldots\ldots\ldots\ldots(8)$$

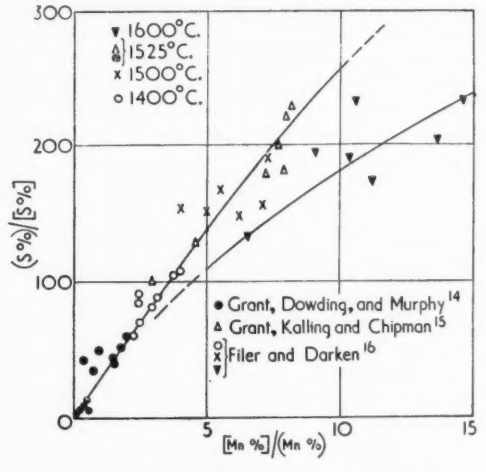

Fig. 3—Sulphur distribution ratio related to inverse of manganese distribution ratio under ironmaking conditions

where ν is a constant for a given value of λ. Values of ν calculated from the log $C_S/(1/T)$ lines at various levels of sulphur capacity index are:

λ	100	60	40	20	
ν	0·084	0·063	0·036	−0·022	
λ	10	8	6	5	4
ν	−0·137	−0·196	−0·287	−0·376	−0·500

The values of ν vary continuously with λ and the equation

$$\nu = -\frac{2{\cdot}68}{\lambda + 0{\cdot}4} + 0{\cdot}11 \quad \ldots\ldots\ldots\ldots(9)$$

is accurate for values of λ less than 20, which cover the entire range of slag compositions of basic steelmaking practice. At higher values of λ the equation is slightly in error.

The sulphur capacity may therefore be expressed as a function of temperature and slag composition in the following form:

$$\log C_S = -\frac{3380}{T} - \frac{2{\cdot}68}{\lambda + 0{\cdot}4} + 0{\cdot}11 \quad \ldots\ldots\ldots(10)$$

Ironmaking Conditions

It is clear that equation (1) could be derived from equation (2) by the addition of equations representing the solution of gaseous sulphur and oxygen in iron. Equation (1) should therefore be valid for conditions other than basic slags and in fact should hold in all basic and acid steelmaking processes or even in a non-ferrous procedure. The applicability of this equation to ironmaking is limited by the fact that the sulphur capacities of the slags are low ($\lambda < 3{\cdot}0$) and that it is extremely difficult to ascertain the oxygen activity in the metal. In view of this, a different method will be sought for formulating the equilibrium sulphur partition between slag and metal under ironmaking conditions.

Grant, Dowding, and Murphy[14] have shown that the sulphur distribution in blast-furnace practice can be related to the manganese distribution. As shown in Fig. 3, this relationship has been extended to an [Mn%]/(Mn%) ratio of about 15 by including the results obtained by Grant, Kalling, and Chipman[15] and Filer and Darken.[16]

Within the temperature range 1400–1500° C., the scatter of experimental points conceals any temperature dependence. When the [Mn%]/(Mn%) ratio is less than 5·0, a straight line may be drawn through

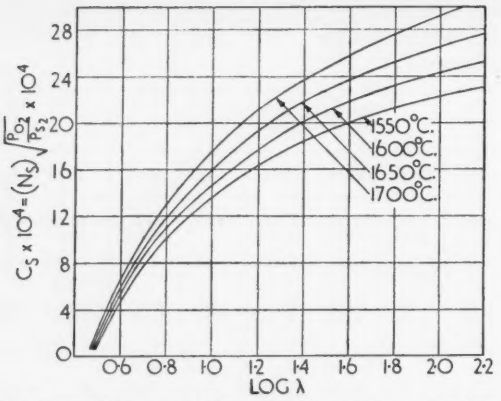

Fig. 4—Sulphide capacity related to slag composition given by sulphur capacity index

the experimental points and the relationship represented by the following simple equation:

$$\frac{(S\%)}{[S\%]} = 28\frac{[Mn\%]}{(Mn\%)} \quad \ldots\ldots(11)$$

This equation meets the requirements of normal blast-furnace practice.

As indicated by Fig. 3, at higher values of metal–slag manganese ratio the temperature effect becomes more prominent.

In view of the fact that the slag and metal compositions studied lie within a narrow range, representing typical blast-furnace conditions, the weight-percentages, instead of ion fractions or activities, were used in calculating the sulphur and manganese slag–metal distribution factors.

EQUILIBRIUM SULPHUR PARTITION BETWEEN GAS AND LIQUID SLAG

As already pointed out, Richardson and Fincham[7, 8] studied the distribution of sulphur and oxygen between gas and liquid slags. Their experiments were, however, limited to simple systems, *e.g.*, calcium silicates, calcium aluminates, magnesium silicate, and calcium aluminosilicates and did not indicate the sulphide capacities of complex slags. An attempt will therefore be made to calculate these from existing data.

The free energy of solution of gaseous sulphur in liquid iron at 1% activity is, according to Sherman, Elvander, and Chipman,[17]

$$\tfrac{1}{2}S_2(g) = [S] \quad \ldots\ldots(12)$$

$$\Delta G^\circ = -31{,}520 + 5{\cdot}27T \quad \ldots(13)$$

Therefore $$\log K_S = +(6890/T) - 1{\cdot}152 \ldots(14)$$

where $$K_S = [a_S]/\sqrt{p_{S_2}} \quad \ldots\ldots(15)$$

Combination of experimental data from the work of Darken and Gurry[18] and Dastur and Chipman[19] yields the following free energy of solution of gaseous oxygen in liquid iron at 1% activity:

$$\tfrac{1}{2}O_2(g) = [O] \quad \ldots\ldots(16)$$

$$\Delta G^\circ = -28{,}220 - 0{\cdot}57T \ldots\ldots(17)$$

and $$\log K_O = +(6170/T) + 0{\cdot}124 \ldots(18)$$

where $$K_O = [a_O]/\sqrt{p_{O_2}} \quad \ldots\ldots(19)$$

Combining equations (4), (15), and (19) yields the sulphide capacity C_S:

$$C_S = (N_S)\sqrt{p_{O_2}/p_{S_2}} \quad \ldots\ldots(20)$$

and summing equations (8), (14), and (18) gives

$$\log C_S = -(2660/T) - 1{\cdot}276 + \nu \ldots\ldots(21)$$

For slags where λ is between 3 and 20 the sum of equations (9) and (21) gives

$$\log C_S = -\frac{2660}{T} - \frac{2{\cdot}68}{\lambda + 0{\cdot}4} - 1{\cdot}166 \ldots\ldots(22)$$

From equation (22) and the values of λ on p. 149 the gas–slag sulphur equilibrium may be represented graphically as in Fig. 4. For slag compositions where $\lambda < 3{\cdot}0$ the value of C_S is very low. The influence of temperature on C_S is in agreement with the findings of Richardson and Fincham,[7, 8] *i.e.*, for a given slag composition, sulphide capacity increases with temperature.

WORKS' DATA

In the light of the present calculations, based on the results of numerous laboratory experiments, it is now possible to find out whether the sulphur distribution between metal and slag actually reaches equilibrium under iron and steelmaking conditions.

Basic Open-Hearth Furnace Data

Works' data, including temperature measurements and complete slag–metal analyses on samples taken during refining and at tapping, were obtained from Stewarts and Lloyds Ltd., John Summers and Sons Ltd., the United Steel Companies Ltd. (these data have been referred to elsewhere by Herasymenko and Speight[20]), and Westfallenhütte, Dortmund (Germany).

To evaluate sulphur capacity it is necessary to know the oxygen activity in the metal in addition to the sulphur distribution ratio. This activity can be readily calculated from the slag composition.[11]

From the known values of λ, (N_S), and $[a_O]$, the sulphur activity in iron can be calculated from equations (4) and (10). The sulphur content of iron can then be found from the equation

$$[S\%] = [a_S]/f_S \quad \ldots\ldots(23)$$

where f_S is the activity coefficient of sulphur dissolved in iron containing various alloying elements. The value of f_S is known for a number of dilute solutions. As pointed out by Wagner,[21] in multi-component systems, where the concentrations of dissolved elements are small, the following equation may be written:

$$f_S = f_S^S \cdot f_S^C \cdot f_S^{Si} \cdot f_S^P \cdot f_S^{Mn} \quad \ldots\ldots(24)$$

where f_S^S, f_S^C, f_S^{Si}, etc., are the sulphur activity coefficient factors in molten Fe–S, Fe–C–S, Fe–Si–S, etc., systems respectively. The experimental results indicate that in dilute solutions the $\log f_S^X/[X\%]$ relationship is linear and not affected by the temperature of the

Table II

INTERACTION PARAMETERS IN Fe–*X*–S MELTS

System	e_S^X	Reference
Fe–S	−0·030	15
Fe–C–S	+0·10	22
Fe–Si–S	+0·075	22
Fe–P–S	+0·040	23
Fe–Mn–S	−0·026	24

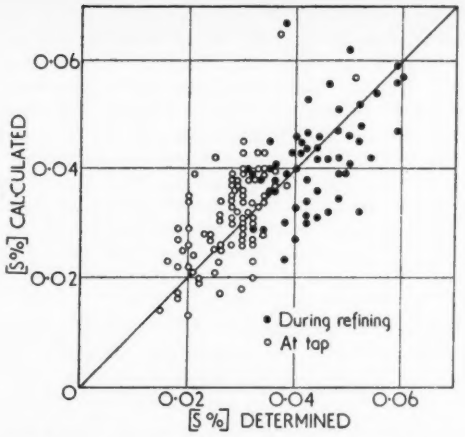

Fig. 5—Relationship between calculated and determined sulphur contents of steel made by basic O.H. processes

melt to a measurable degree. In Table II the values of interaction parameter, *i.e.*,

$$e_S^X = \frac{\partial \log f_S^X}{\partial [X\%]} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (25)$$

are given for a number of Fe–*X*–S melts, together with references from which the data have been taken.

The proportion of sulphur dissolved in iron, in equilibrium with a slag of known composition, can therefore be calculated, and in Fig. 5 the calculated sulphur content of the metal is plotted against the sulphur determined by analysis. The points lie evenly about a line drawn through the origin of the co-ordinates with a slope of unity. The band, within which 70% of the points are located, corresponds to about $\pm$ 0·008% S. It appears that during refining the transfer of sulphur between the slag and metal phases is at equilibrium and that equilibrium is restored shortly after any disturbance due to bath or slag additions or to temperature changes.

In the data given by Herasymenko and Speight,[20] the slags contained small proportions of fluoride, and therefore due allowance was made for the effect of fluoride ions on the ferrous oxide activity.[25] No correction was, however, necessary to the value of $\mathfrak{S}_S$, because as already pointed out sulphur capacity is not influenced by fluoride ions.

The points in Fig. 5 are based on data from four sources, and further consideration is being given to the universal application of the information given in this diagram.

Blast-Furnace Data

The application of physico-chemical concepts to blast-furnace practice is often made difficult by an inability to obtain temperatures and slag and metal samples which are representative of the conditions in the hearth.

An attempt will, nevertheless, be made here to show whether there is any tendency for the slag and metal to reach a state of equilibrium with respect to the sulphur transfer between the two phases. There are two sources of works' data available:

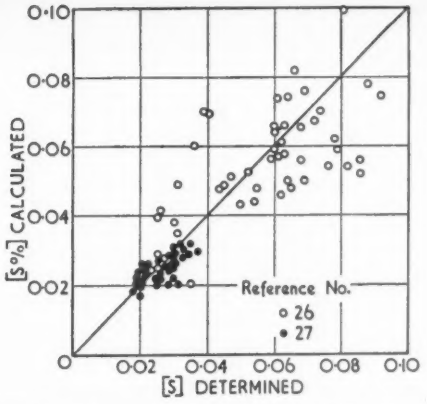

Fig. 6—Relationship between calculated and determined sulphur contents of pig iron

(i) The B.I.S.R.A. report[26] of a survey of blast-furnace practice in which data relating to 59 furnaces are given, derived from 70% of the blast-furnace plants in Great Britain for the period 1930–47

(ii) The monthly average slag and metal analyses for the years 1948, 1949, and 1950, of the Ford Motor Co., Ltd., Dagenham.[27]

In no case was the temperature measured at the time of sampling.

When the manganese and sulphur contents of the slag and the manganese content of the metal are known, the concentration of sulphur in iron in equilibrium with the slag, and the sulphur partition, can be calculated from equation (11). In Figs. 6 and 7 the calculated sulphur dissolved in iron and the calculated sulphur distribution ratio are plotted against those determined by analysis. Considering that the data used were averages only, the scatter of the points about the lines drawn is relatively small, indicating that the sulphur partition between metal and slag reaches equilibrium in the blast-furnace at the time of tapping.

DISCUSSION

It has been shown in Figs. 1*a–d* that the sulphur capacity $\mathfrak{S}_S$ can be related to a λ term which is a

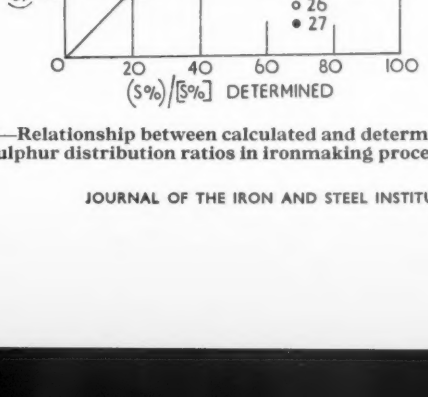

Fig. 7—Relationship between calculated and determined sulphur distribution ratios in ironmaking processes

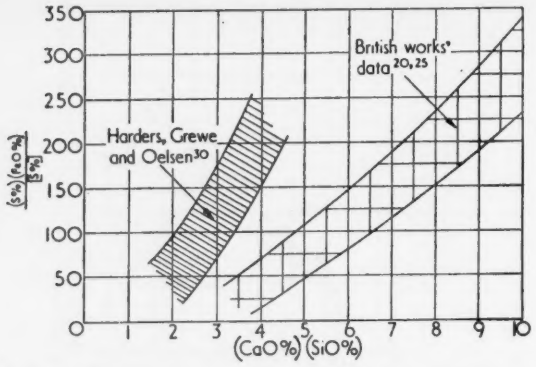

Fig. 8—Relationship between (S%)(FeO%)/[S%] and (CaO%)/(SiO_2%) from works' basic O.H. data

function of the oxygen-ion activity of the slag. Since the latter cannot yet be determined experimentally, nor can it be calculated on any theoretical grounds, several attempts have in the past been made to find a substitute for the oxygen-ion activity by assuming, for example, (i) the existence of 'free' and 'combined' oxides in slags,[1] (ii) the formation of neutral slags, which allows the calculation of 'excess base' and similar terms,[3, 28, 29] and (iii) the complete ionization of molten slags to certain types of ions.[5] Although the application of such concepts to slag–metal equilibrium data may be satisfactory within a limited composition range, or their use may be imperative through lack of adequate knowledge of the true state of the reacting entities in slags, great care should be observed in the employment of empirical methods, because they often lead to misinterpretation of the experimental data and concealment of some important facts. For example, although the method chosen by Grant and Chipman[3] for relating the sulphur distribution factor (S%)/[S%] to slag composition elucidated certain aspects of the sulphur reaction, it overlooked the important influence of the oxygen potential at the slag/metal interface on the sulphur partition and made the effect of temperature difficult to clarify.[29]

The choice of equation (7) for the evaluation of sulphur capacity index is based on the following reasons. The slag components which form oxyacid radicles are silicon, phosphorus, and aluminium. In basic slags silica, phosphorus pentoxide, and alumina can be assumed to dissolve in the slag as SiO_4^{4-}, PO_4^{3-}, and AlO_3^{3-}. That is,

$$SiO_2 + 2(O^{2-}) = (SiO_4^{4-}) \quad \ldots\ldots(26)$$

$$P_2O_5 + 3(O^{2-}) = 2(PO_4^{3-}) \quad \ldots\ldots(27)$$

$$Al_2O_3 + 3(O^{2-}) = 2(AlO_3^{3-}) \quad \ldots\ldots(28)$$

In other words, each molecule of phosphorus pentoxide or of alumina will combine with 1·5 times as many oxygen ions as will silica. It should, however, be pointed out that alumina is an amphoteric oxide. For example, in slags of low basicity, *i.e.*, low oxygen-ion activity, alumina can behave like a base as a result of the dissociation of AlO_3^{3-} radicles to Al^{3+} cations and oxygen ions, thus

$$(AlO_3^{3-}) = Al^{3+} + 3(O^{2-}) \quad \ldots\ldots(29)$$

This is why equation (7) cannot be applied to calcium aluminosilicate slags which contain high proportions of alumina similar to those used by Richardson and Fincham[7] in their experimental work. On the other hand, equations (28) and (29) indicate that in basic slags, where the oxygen-ion activity is higher, there will be a strong tendency for the formation of AlO_3^{3-} oxyacid radicles. The dissociation of SiO_4^{4-} and PO_4^{3-} radicles will again be influenced by the oxygen-ion activity of the slag, and this will give rise to the formation of Si–O and P–O polymers as the slag becomes richer in acidic oxides.

From these arguments it follows that equation (7) is only a measure of 'oxygen-ion activity or concentration' of the slag, where the proportion of alumina is less than about 0·15 mole fraction (*see* Table I).

Another interesting fact to be observed from $\mathfrak{L}_S$ *vs.* λ and C_S *vs.* λ relationships is that the nature of the cations does not appear to affect the sulphur capacity. Although the cations do not participate in reactions (1) and (2), they should influence the activities of sulphur and oxygen ions in the slag. In fact Richardson and Fincham have shown that for a simple slag of given basicity, the sulphide capacity varies with the nature of the cation. It appears therefore that in complex slags, the influence of the various cations on the ratio of the activity coefficients of oxygen and sulphur ions is not detectable. The equivalence of calcium oxide, magnesium oxide, and manganous oxide on the activity of ferrous oxide[11] is probably an analogous phenomenon.

Although the calculations based on basic open-hearth data indicate that the partition of sulphur between slag and metal phases attains equilibrium during refining, attention should be drawn to the possibility of sulphur pick-up by the slag. For instance, the sulphur content of the steel bath calculated from the works' data given by Harders, Grewe, and Oelsen[30] was found to be lower than that by analysis. In other words, the sulphur contents of the slags examined by them are high compared with those from laboratory experiments, where the slag–metal equilibrium was established intentionally, and those from the works' data used in this paper. To illustrate this anomaly, Fig. 8 has been drawn in the manner adopted by Harders, Grewe, and Oelsen. It will be seen that for a given CaO/SiO_2 ratio and ferrous oxide concentration, the sulphur partition is much higher in the one case than in the others. Assuming that the sulphur analysis in iron was correct, the sulphur in the slag was higher than it should have been. If on the other hand their data are accepted as correct, this would mean that the iron was desulphurized by the slag to an extent greater than equilibrium would allow, which is not possible. It appears, therefore, that either the rate of sulphur pick-up by the slag was greater than the rate of transfer of sulphur from slag to metal, or that the slag picked up sulphur from the furnace gases during the process of sampling. The latter is believed to be more probable than the former; slags ought therefore to be sampled in a manner which will prevent the contamination of the sample by iron oxide dust and sulphur from the furnace atmosphere. This

may be accomplished by employing the 'bomb' mould which is often used in taking metal samples.

Although the relationship in Fig. 8 bears a similarity to those in Figs. 1*a* and 1*d*, the former is only qualitative and cannot be used in predicting the correct equilibrium conditions with respect to the sulphur reaction.

The calculations based on the British works' data also indicate that the oxygen activity in iron, derived from the ferrous oxide iso-activity diagram,[11] does in fact represent the oxygen potential at the slag/metal interface.

In dealing with the equilibrium sulphur partition between slag and metal under reducing conditions, *e.g.*, blast-furnace practice, the manganese distribution ratio appears to be of great importance, and the relationship in Fig. 3 is to be expected for the following reasons. The oxidation of manganese and the dissolution of the reaction product in the slag may be represented by the equations

$$[\text{Mn}] + [\text{O}] = \text{MnO} \quad \ldots\ldots(30)$$

and

$$\text{MnO} = (\text{Mn}) + (\text{O}) \quad \ldots\ldots(31)$$

Combination of these reactions with reaction (1) yields

$$\underset{\text{metal}}{[\text{Mn}]} + [\text{S}] = (\text{Mn}) + \underset{\text{slag}}{(\text{S})} \quad \ldots\ldots(32)$$

for which it follows that

$$\frac{(a_{\text{S}})}{[a_{\text{S}}]} = K\frac{[a_{\text{Mn}}]}{(a_{\text{Mn}})} \quad \ldots\ldots(33)$$

Since the plot of (S%)/[S%] against [Mn%]/(Mn%) in Fig. 3 gives a curve it follows that the $(\gamma_{\text{Mn}})(\gamma_{\text{S}})/[\gamma_{\text{Mn}}][\gamma_{\text{S}}]$ term changes regularly with the manganese distribution ratio. That is, the product of the activity coefficient terms mentioned and the slope of the isothermal in Fig. 3 will give the equilibrium constant K in equation (33).

The reaction

$$\underset{\text{metal}}{[\text{Fe}]} + [\text{S}] = (\text{Fe}) + \underset{\text{slag}}{(\text{S})} \quad \ldots\ldots(34)$$

must also occur in the blast-furnace but attempts to evaluate the equilibrium constant have been unsatisfactory because of the difficulty experienced in separating the metallic iron from blast-furnace slags.[29] This very fine metallic iron dispersed in the slag arises from the reduction of dissolved iron oxide in slag by suspended carbon.

The sulphur reaction in blast-furnace practice has recently been subject to intensive studies. It has been shown[14, 15] that additions of manganous oxide, silica, and ferrous oxide to a slag in equilibrium with a carbon-saturated iron causes an almost immediate sulphur transfer to the metal from the slag layer as a result of an increase in the oxygen potential at the slag/metal interface, in spite of the fact that the metal is saturated with carbon. These workers[14, 15] have not, however, considered the important role of the kinetics of the carbon–oxygen reaction on the slag–metal sulphur distribution.

Filer and Darken[16] remelted blast-furnace metal and slag in a small graphite crucible and found that the sulphur content of metal was reduced; there were also reversions of manganese and silicon from slag to the metal. Their conclusion was that in a blast-furnace the distributions of silicon, manganese, and sulphur between slag and metal did not reach equilibrium. Supporting this view Grant, Dowding, and Murphy[14] plotted the sulphur distribution ratio against 'excess CaO,' as defined by Grant, Kalling, and Chipman[15]; according to this relationship, for a given 'excess CaO' the sulphur distribution ratio is higher in laboratory melts than those found from blast-furnace slag and metal samples. The main objection to this argument is that the oxygen activity or content in iron was not taken into account. On the other hand, the application of the relationship in Fig. 3 or equation (11) to works' data indicates (*see* Figs. 6 and 7) that there is good agreement between the calculated and determined values of [S%] and (S%)/[S%], which supports the view that in a blast-furnace an equilibrium is established between slag and metal with respect to the sulphur reaction. A similar argument holds for the manganese reaction.

In view of the present evidence, it is believed that in blast-furnace practice the carbon–oxygen reaction in iron is sluggish, *i.e.*, the oxygen content of the carbon-saturated iron is much greater than has hitherto been assumed, and consequently, on remelting blast-furnace iron and slag in a graphite crucible with efficient stirring, the excess dissolved oxygen is gradually reduced. As a result, the following reactions take place concurrently: (i) return of silicon to the metal, (ii) return of manganese to the metal, and (iii) removal of sulphur from the metal. In other words, on remelting, the state of slag–metal equilibrium with respect to manganese, sulphur, and silicon reaction is moved from one position to another through the deoxidation of the metal by carbon. It appears that the sluggishness of the carbon–oxygen reaction noted in open-hearth and electric-arc processes[25] is also evident at much higher carbon concentrations.

To avoid any misinterpretations of the present arguments it should be stated that under iron and steelmaking conditions there are at least four or more phases, *i.e.*, gas, 'hearth and furnace walls,' slag, and metal; because of various factors affecting the reaction rates, it is never possible to obtain a complete state of equilibrium in these complex systems, but this does not prevent the establishment of equilibrium between some phases with respect to certain reacting entities. If the rate of a sluggish reaction in this system is increased, *e.g.*, by remelting blast-furnace iron and slag in a graphite crucible, the state of equilibrium which previously existed with respect to other slag–metal components will be disturbed but re-established at a different level.

CONCLUSIONS

The sulphur reaction in slag–metal systems can be represented by

$$[\text{S}] + (\text{O}) = [\text{O}] + (\text{S})$$

The sulphur capacity, *i.e.*,

$$C_{\text{S}} = (N_{\text{S}})\,[a_{\text{O}}/a_{\text{S}}]$$

is a measurable quantity and is related to a sulphur capacity index

$$\lambda = 1/(N_{\text{SiO}_2} + 1{\cdot}5N_{\text{P}_2\text{O}_5} + 1{\cdot}5N_{\text{Al}_2\text{O}_3})$$

The sulphur capacity index is a measure of the oxygen-ion activity of the slag.

The sulphur capacity of a slag depends on composition and temperature:

$$\log \mathfrak{K}_S = -(3380/T) + \nu$$

where $$\nu = -\frac{2 \cdot 68}{\lambda + 0 \cdot 4} + 0 \cdot 11$$

when $\lambda < 20$.

For ironmaking conditions the sulphur distribution can be related to the manganese distribution. When the [Mn%]/(Mn%) ratio is less than 5·0 and the temperature is within the 1400–1500° C. range the following equation applies:

$$(S\%)/[S\%] = 28[Mn\%]/(Mn\%)$$

For a given manganese distribution ratio, the sulphur distribution ratio decreases with rising temperature.

The sulphur reaction in gas–slag systems is represented by

$$\tfrac{1}{2}S_2(g) + (O) = \tfrac{1}{2}O_2(g) + (S)$$

and sulphide capacity, *i.e.*,

$$C_S = (N_S)\sqrt{p_{O_2}/p_{S_2}}$$

depends on temperature and composition in the following way:

$$\log C_S = -(2660/T) - 1 \cdot 276 + \nu$$

Applying these findings to works' data indicates that the sulphur transfer to and from the slag and metal phases attains equilibrium under normal iron and steelmaking conditions.

One of the most important conditions that affect the sulphur reaction is the oxygen potential of the system at the slag/metal interface. The calculations involving oxygen activity or concentration should not be based on the carbon concentration in the metal nor should the oxygen content in iron, obtained by direct analysis, be used if the concentration of carbon is greater than 0·05%. This has been discussed in more detail elsewhere.[25]

The belief that the sluggishness of the carbon–oxygen reaction in iron is the cause of the apparently high residual sulphur in blast-furnace iron has been confirmed.[31] Despite this, however, in blast-furnace practice the sulphur and manganese distributions between slag and metal attain equilibrium. This state of equilibrium can be changed by removing the excess dissolved oxygen in carbon-saturated iron either by allowing the carbon–oxygen reaction to proceed further or by adding a suitable deoxidizer, *e.g.*, aluminium, silicon, manganese, etc.; this is followed by the return of silicon and manganese to metal and a lowering of the sulphur content of the iron.

Acknowledgments

The author wishes to thank Dr. J. Pearson for helpful discussions, and Stewarts and Lloyds Ltd., John Summers and Sons Ltd., and the Ford Motor Co., Ltd., for their works' data. In the preparation of this paper great use has been made of the data produced by J. Chipman and co-workers from their laboratory slag–metal equilibrium measurements.

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Thermodynamics of Carbon Dissolved in Iron Alloys

Part III: SOLUBILITY OF CARBON IN IRON–SULPHUR MELTS

By E. T. Turkdogan, Ph.D., M.Met., and R. A. Hancock, B.Sc.

SYNOPSIS

The solubility of carbon in iron–sulphur melts has been determined at 1200°, 1350°, and 1500° C. Dissolved sulphur decreases the solubility of carbon to a greater extent than do silicon and phosphorus. The miscibility gap at carbon saturation has also been investigated at the same temperatures, and the results are compared with those of earlier workers. A tentative method has been formulated for calculating the activity coefficient of carbon in iron–sulphur–carbon ternary melts. 1053

DURING recent work[1] on the influence of silicon, manganese, and aluminium on the rate of desulphurization of carbon-saturated iron, samples of molten iron containing only sulphur and carbon were analysed. The results indicated that at a given sulphur concentration and temperature the carbon saturation was approximately 0·5% higher than that found by Kitchener, Bockris, and Spratt.[2] In view of this significant difference, it was felt desirable to reinvestigate in detail the influence of sulphur on the solubility of graphite in iron.

EXPERIMENTAL WORK

Apparatus

The furnace and its auxiliary equipment were similar to those described in Part I of this series of papers[3] and they will not be described here.

For the determination of the miscibility gap in the iron–sulphur–carbon system at carbon saturation, a graphite crucible of ½ in. dia. and 9 in. long and supported by a stainless-steel tube from the top rubber bung, was used. Suitably chosen amounts of an iron–sulphur–carbon alloy and ferrous sulphide were packed into the bottom inch of the crucible and the whole assembly was lowered into the reaction tube, so that the charge was in the middle of the hot zone.

Materials

The master alloy, about 120 g. in weight, consisted of a specially prepared iron–sulphur–carbon alloy containing 1·4% of sulphur and 4% of carbon. To obtain charges with much lower sulphur contents, the master alloy was diluted with B.I.S.R.A. *AG* iron, the composition of which is given elsewhere.[3]

Procedure

A graphite crucible was charged with the master alloy and was supported in the reaction tube in which a flow of argon maintained an inert atmosphere. When molten, the contents of the crucible were kept at a fixed temperature until equilibrium was attained. The melt was then sampled with a silica tube and evacuated bulb assembly[3] and the sample was analysed for carbon and sulphur. After the withdrawal of the sample, sufficient B.I.S.R.A. *AG* iron was added to the melt in small pieces to lower the sulphur content by about 0·2%. Sufficient time was allowed to elapse for the melt to rehomogenize and for the carbon content to reach its equilibrium value. The melt was again sampled and analysed. This procedure was repeated at a fixed temperature to give samples having sulphur contents ranging from 1·4 to 0·03%.

This procedure was carried out at 1200°, 1350°, and 1500° C. For each temperature a new graphite crucible containing a fresh charge was used.

Earlier experiments[3] had indicated that iron–phosphorus melts required 1 hr. to reach carbon saturation equilibrium. In the present work additional experiments were carried out to ascertain that the equilibrium had been established at the time of sampling. After sample *K*1 was taken about 20 g. of *AG* iron was added to the melt to reduce its sulphur content by about 0·2%; an hour after this addition sample *K*2 was taken and then the melt was kept undisturbed at 1200° C. for another hour and sample *K*3 was taken. The compositions of the samples are identical within the limits of analytical accuracy. It is also interesting to note that the loss of sulphur due to prolonged heating in argon is negligible. In the subsequent experiments at least 1½ hr. were allowed between addition of the pure iron and taking of the sample.

The above procedure ensured that equilibrium was

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Table I

SOLUBILITY OF CARBON IN IRON–SULPHUR MELTS

No.	Temp., °C.	Weight-percentage		Atom Fraction	
		S	C	S	C
*K*1	1200	1·175	4·089	0·0177	0·1643
*K*2	,,	1·073	4·173	0·0161	0·1669
*K*3	,,	1·065	4·176	0·0160	0·1674
*K*4	,,	0·786	4·213	0·0118	0·1680
*K*5	,,	0·354	4·306	0·0053	0·1726
*K*6	,,	0·272	4·278	0·0040	0·1720
*L*1	1350	1·430	4·430	0·0213	0·1759
*L*2	,,	1·110	4·480	0·0165	0·1779
*L*3	,,	0·883	4·535	0·0131	0·1801
*L*4	,,	0·656	4·520	0·0098	0·1798
*L*5	,,	0·366	4·690	0·0054	0·1859
*L*6	,,	0·244	4·720	0·0036	0·1872
*M*1	1500	1·210	4·865	0·0178	0·1908
*M*2	,,	0·980	4·900	0·0144	0·1922
*M*3	,,	0·833	4·830	0·0123	0·1901
*M*4	,,	0·760	4·900	0·0112	0·1925
*M*5	,,	0·630	4·960	0·0093	0·1947
*M*6	,,	0·574	4·959	0·0084	0·1947
*M*7	,,	0·550	4·940	0·0081	0·1941
*M*8	,,	0·542	4·969	0·0080	0·1950
*M*9	,,	0·497	5·044	0·0073	0·1976
*M*10	,,	0·360	5·035	0·0053	0·1974
*M*11	,,	0·270	5·130	0·0039	0·2007
*M*12	,,	0·089	5·065	0·0010	0·1990
*M*13	,,	0·042	5·212	0·0006	0·2030
*M*14	,,	0·027	5·261	0·0004	0·2050

approached from a lower carbon level and that supersaturation was prevented.

The miscibility experiments were carried out in the apparatus described, and the temperature of the melt was estimated by means of a thermocouple in a mullite sheath placed alongside the crucible in the reaction tube. The temperature of the melt was maintained constant for at least 4 hr. and the stainless-steel supporting tube and graphite crucible were then removed from the furnace and plunged into water; this operation took about 10 sec. On breaking open the crucible, it was found that the charge had separated into two distinct layers which could easily be detached by gently tapping at the phase boundary with a cold chisel. Figure 1 is a photograph of one of the samples which consisted of two well-separated phases. In each case the two layers were carefully cleaned and the surface of the carbon-rich phase, *i.e.*, the bottom layer, ground off before crushing and analysing for carbon and sulphur.

Analysis

The carbon contents of the samples were estimated by the combustion method, agreement between duplicates being within $\pm 0{\cdot}01\%$. In the case of sulphur-rich miscibility samples, extra absorption bulbs containing chromic acid were placed in the gas train to strip out the large volume of sulphur dioxide produced on combustion of the samples.

The sulphur contents were determined by the combustion method using a Ridsdale white cast iron to standardize the sodium borate solution. One-volume hydrogen peroxide was used as absorbent. Agreement between duplicates was within $\pm 0{\cdot}02\%$ except in the case of the sulphur-rich miscibility samples, which showed a slightly greater divergence. For the latter samples, the material was mixed with a weighed quantity of the standard white cast iron to obtain a steadier combustion.

RESULTS

Solubility of Carbon in Iron–Sulphur Melts

The effect of sulphur on the solubility of graphite in iron has been investigated at 1200°, 1350°, and 1500° C. and at sulphur concentrations up to the miscibility limit. The experimental results, excluding those relating to miscibility, are given in Table I.

As indicated by Fig. 2, where the atom fraction of carbon at saturation is plotted against the atom fraction of sulphur, the solubility of carbon in iron is reduced by dissolved sulphur. The shape of the curves is similar to those found for iron–phosphorus–carbon ternary melts.[3] The isothermals are terminated at points where the melts separate into two layers.

Miscibility Gap in Carbon-Saturated Iron–Sulphur–Carbon Melts

The miscibility gap in the iron–sulphur–carbon system had been determined several times; reference may be made to Schenck[4] for a review of earlier work. In the present investigation it was decided to determine the boundaries of the miscibility gap at 1200°, 1350°, and 1500° C. and the experimental results are given in Table II. Figure 3 shows the relationship between the temperature and composition of the carbon-rich layer, *i.e.*, the bottom layer; the results of earlier work are also included. There is a good agreement between the present authors' values and those suggested by Darken and Larsen[5] who stated that 'the influence of carbon upon the activity of sulphur may be estimated from some hitherto unpublished data obtained here in association with R. W. Gurry, which indicate that at 1500° C. in an atmosphere of nitrogen the solubility of FeS in iron saturated with graphite (about 4% carbon) corresponds to a sulphur content of about 1·5%.' The sulphur contents of the carbon-rich layer, determined by Norrö and Lundquist,[6] are much higher than those

Table II

MISCIBILITY GAP IN CARBON-SATURATED IRON–SULPHUR–CARBON MELTS

No.	Temp., °C.	Weight-percentage	
		S	C
Carbon-rich layer **(*bottom layer*)**			
*N*1	1200	1·82	3·60
*N*2	1350	1·94	3·78
*N*3	1500	1·77	4·23
Sulphur-rich layer **(*top layer*)**			
*O*1	1200	28·4	0·35
*O*2	1350	26·9	0·72
*O*3	1500	26·5	0·90

Fig. 1—The two layers in the iron–sulphur–carbon system

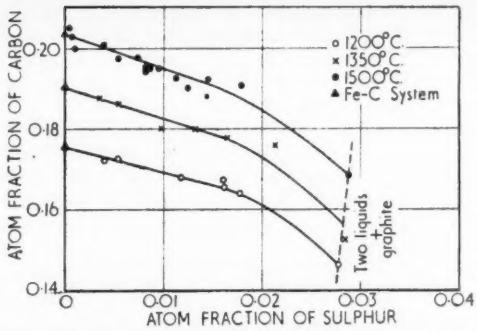

Fig. 2—Effect of sulphur on solubility of graphite in molten iron

found by other workers. It is believed that the method of sampling adopted by Norrö and Lundquist did not prevent the contamination of the bottom-layer sample with the sulphur-rich top layer. These authors did not analyse bottom-layer samples for carbon content but calculated them by an approximate method; these values, which may not be correct, are not included in Fig. 3. Hanemann and Schildkötter[7] determined the miscibility gap in iron–sulphur–carbon ternary melts in magnesia crucibles at lower carbon concentrations. They found that in this system there is a 4-phase plane at 1105° C. at which two liquids are in equilibrium with cementite (or graphite) and austenite. The composition of the carbon-rich layer monotectic point is given in Fig. 3. The sulphur content of the alloy at this temperature is much lower than that estimated by the extrapolation of the sulphur solubility curve.

Figure 4 shows the relationship between the temperature and composition of the sulphur-rich layer, *i.e.*, the top layer. There appears in this case to be better agreement amongst the various workers' results.

DISCUSSION

It was shown by Turkdogan and Leake[3] that ΔN_C^X was a temperature-independent function of the atom fraction of X, where X represents silicon, phosphorus, and manganese. The term ΔN_C^X is the difference between the carbon saturation in an iron–X–carbon

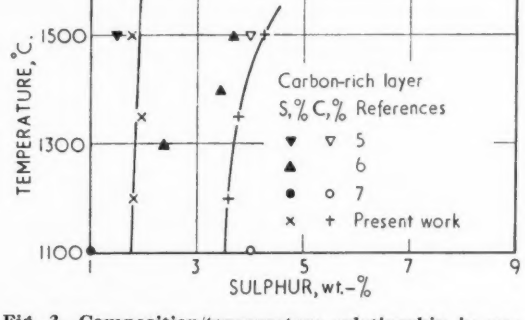

Fig. 3—Composition/temperature relationship in carbon-rich layer of two immiscible liquids in iron–sulphur–carbon melts

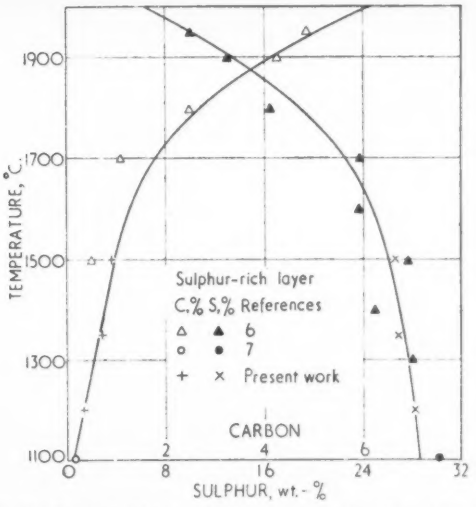

Fig. 4—Composition/temperature relationship in sulphur-rich layer of two immiscible liquids in iron–sulphur–carbon melts

ternary melt and that in an iron–carbon binary melt at the same temperature. The present results and those of Morris and Buehl[8] have been used to plot (Fig. 5) $-\Delta N_C^S$ against the atom fraction of dissolved sulphur. The scatter of the points in Fig. 5 is well within the analytical errors involved in the determinations of high percentages of sulphur and carbon.

The relationship in Fig. 5 is assumed to be linear up to about 0·01 atom fraction of sulphur, *i.e.*, up to about 0·7% of sulphur, and the following equation may be written:

$$\Delta N_C^S = -N_S \text{ (when } N_S < 0{\cdot}01) \quad \ldots\ldots\ldots\ldots (1)$$

where

$$\Delta N_C^S = N_C - N_C^C \quad \ldots\ldots\ldots\ldots\ldots\ldots (2)$$

N_C^C and N_C are the atom fractions of carbon in iron–carbon and iron–sulphur–carbon melts, respectively, and N_S is the atom fraction of sulphur in the ternary melts. The value of N_C^C in carbon-saturated melts can be calculated at any temperature from 1153° (eutectic temperature) to 2000° C. by the following equation[3]:

$$\log N_C^C = -(560/T) - 0{\cdot}375 \quad \ldots\ldots\ldots\ldots (3)$$

Values of ΔN_C^S and N_S have been calculated from

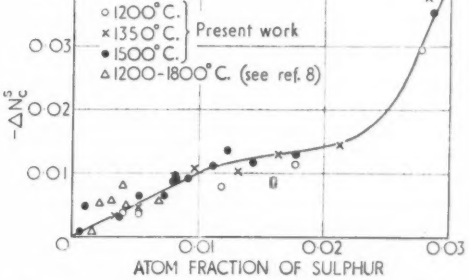

Fig. 5—Relationship between $-\Delta N_C^S$ and sulphur concentration

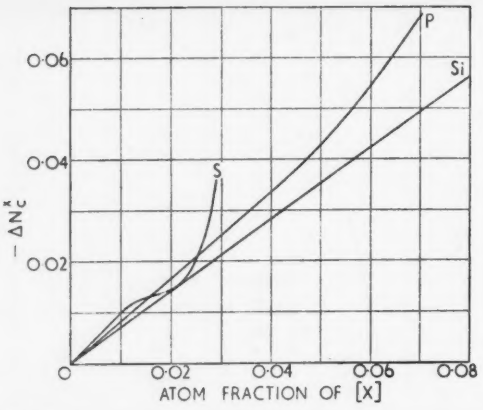

Fig. 6—Comparison of effects of sulphur, phosphorus, and silicon on solubility of graphite in molten iron. This relationship is independent of temperature

the results of Kitchener, Bockris, and Spratt[2] and plotted as in Fig. 5; the slope of a line drawn through their points is three times that given by equation (1). Moreover, the scatter of the points about the mean line is five times that found in Fig. 5. It is not clear why the carbon saturation obtained by Kitchener, Bockris, and Spratt should be about 0·5% lower than those of the present authors. The supersaturation of iron–sulphur–carbon melts with respect to carbon is a well-recognized phenomenon which is associated with the stable nature of cementite in the presence of dissolved sulphur. The concordant results obtained from the present work, however (*see* Figs. 2 and 5), indicate that the melts were free of carbon supersaturation.

In Part I of this series[3] attempts were made to indicate the causes of the observed effects of alloying elements on the solubility of graphite in iron on the basis of

(i) Solute–solute and solute–solvent interactions
(ii) Ionization of atoms when dissolved in iron
(iii) The size-factor, which is determined by the ionic radii of solvent and solute
(iv) The structural difference between iron–X and iron–carbon melts.

The observed changes in the thermodynamics of carbon in iron, brought about by the presence of an alloying element, are due to the relative magnitudes of these four effects.

In Fig. 6, $-\Delta N_C^X$ is plotted against the atom fraction of X, where X represents sulphur, phosphorus or silicon. It is interesting to note that for a given concentration of element X, the carbon displaced from the melt becomes greater as the element X in the second Short Period becomes more electronegative. Between 0·01 and 0·025 atom fraction of sulphur there is, however, an inflection in the curve which does not appear in the curves for the iron–phosphorus–carbon and iron–silicon–carbon systems.

In the case of iron–sulphur–carbon melts, the major cause of the lowering of the solubility of carbon in iron by sulphur is believed to be the structural differences between iron–sulphur and iron–carbon systems, as pointed out by Kitchener, Bockris, and Spratt.[2] It is mainly for this reason that this ternary melt will separate into two layers and that the miscibility gap extends over a large area, as may be observed from Figs. 3 and 4.

In view of the discrepancy of the results on the composition/temperature boundary of the carbon-rich phase (Fig. 3) it appears desirable to extend this work to temperatures higher than 1500° C. and also to re-investigate the miscibility boundary in the absence of carbon saturation.

It was shown by Turkdogan[9] that the value of $\Delta N_C^{Si}/N_C^C$ as a function of atom fraction of silicon was independent of carbon activity in iron–silicon–carbon alloys at 1000° C. and it was assumed that a similar condition held for iron–silicon–carbon melts. Using this assumption, the activity coefficient of carbon in these ternary melts was computed. It is difficult to predict the extent to which the assumption will hold for iron–sulphur–carbon melts. In the absence of any experimental data on the influence of sulphur on the activity of carbon, occasions may arise when it is desirable to assess, at least approximately, the magnitude of this effect. Combination of equations (1) and (3) yields

$$\log \frac{-\Delta N_C^S}{N_C^C} = \frac{560}{T} + \log N_S + 0{\cdot}375 \quad \ldots\ldots\ldots (4)$$

Equation (4) is based on results at unit carbon activity and, if it holds good at lower carbon activities, it then becomes possible to calculate the activity coefficient of carbon by the method of computation described in Part II of this series.[9]

CONCLUSIONS

Sulphur lowers the solubility of graphite in iron. This effect is best illustrated by the relationship between ΔN_C^S and atom fraction of sulphur; this function is temperature-independent and may be represented by

$$\Delta N_C^S = -N_S$$

when $N_S < 0{\cdot}01$ ($< 0{\cdot}7\%$ of sulphur), *i.e.*, at a given temperature, for every atom of sulphur dissolved in carbon-saturated iron, an atom of carbon is displaced from the solution.

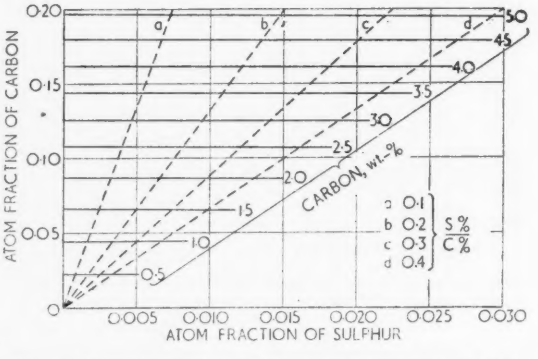

Fig. 7—Interconversion of atom fractions and weight-percentages in iron–sulphur–carbon systems

The sulphur content of the carbon-saturated miscibility boundary has been found to increase only slightly with rise of temperature, *i.e.*, from 1·82% at 1200° C. to 1·92% at 1500° C. This increase is much less than that found by previous workers.

Since there are no experimental data from which the activity coefficient in iron–sulphur–carbon ternary melts may be derived, a method has been suggested for computing them.

Acknowledgment

The authors wish to thank Dr. J. Pearson for the interest he has shown in this work.

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APPENDIX

Graphical Interconversion of Atom Fractions and Weight-Percentages

In Fig. 7 the atom fraction of carbon is plotted against the atom fraction of sulphur for ranges of carbon weight-percentages and ratios S%/C% in iron–sulphur–carbon ternary melts. The point of intersection of the lines gives the atom fractions of sulphur and carbon corresponding to weight-percentages in the ternary system.

A Microscopical Examination of Samples of Iron Containing Siliceous Inclusions

By R. E. Lismer, A.Met., F.I.M., and F. B. Pickering, A.Met., A.I.M.

SYNOPSIS

A detailed microscopical examination has been carried out on samples of iron containing siliceous inclusions, along the lines adopted for the previous investigations on manganous and aluminous inclusions. The object was to establish the means of positive identification by the determination of the optical and chemical properties of the inclusions. 1037

IT was decided by the Solid Steel Study Group that Evans and Sloman's microscopical examination[1] of the inclusions associated with the deoxidation of pure-iron–iron-oxide melts by silicon should be supplemented by details of the optical and chemical properties. This work was to follow the style adopted by Lismer and Pickering [2, 3] and the data obtained were to provide identification methods for some of the siliceous types of inclusion.

Material Examined

Samples for examination were taken from ingots *SA*, *SC*, *SE*, *SG*, *SL*, and *SM* of the series described by Evans and Sloman.[1] The crystalline constituents of the inclusions in each ingot were determined by the X-ray examination of alcoholic iodine residues. The results are given in Table I, which is an abstract of that compiled by Evans and Sloman.

Polishing Technique

The polishing technique and methods of microscopical examination were similar to those described in the previous papers by Lismer and Pickering.[2, 3] There was a tendency for the inclusions in the present series to become pitted very easily and precautions against over-polishing were essential.

Paper SM/C/52/54 of the Physical Chemistry of Steelmaking Committee of the Steelmaking Division of the British Iron and Steel Research Association, received on 1st July, 1954. The views expressed are the authors' and are not necessarily endorsed by the Committee as a body.

Mr. Lismer is in the Research Department of Lloyd's Register of Shipping. Mr. Pickering is in the Research and Development Department of The United Steel Companies Ltd.

Method of Examination

The scheme of examination used the following :

(*a*) Normal vertical illumination with a light etch in alcoholic 2% nitric acid as desired
(*b*) Polarized vertical illumination
(*c*) Whiteley's silver nitrate reagent for sulphides
(*d*) Alcoholic 10% nitric acid
(*e*) Alcoholic 5% hydrochloric acid
(*f*) Aqueous 20% hydrofluoric acid
(*g*) Other etchants as described later.

EXPERIMENTAL RESULTS

Examination under Normal Vertical Illumination

Ingot SA—The inclusions were generally evenly distributed, showing a tendency to form small groups. They were mainly monophased rounded particles of wüstite (FeO), light grey in colour, with some containing an inner metallic particle (Fig. 1).

Ingot SC—The distribution of the inclusions was similar to that in ingot *SA*. The majority were duplex consisting of light-grey wüstite and a darker grey phase, fayalite ($2FeO.SiO_2$). Several predominant forms were observed :

(*a*) A globular type consisting mainly of the fayalite–wüstite eutectic with a little primary wüstite (Fig. 2*a*). Occasionally an outer periphery of fayalite was evident.
(*b*) A globular type containing much primary wüstite in a matrix of fayalite–wüstite eutectic (Fig. 2*b*). Many of the inclusions showed a degenerate eutectic, the wüstite globules being embedded in a fayalite matrix (Fig. 2*c*).
(*c*) Occasional dumb-bell shaped particles of wüstite and fayalite (Fig. 3*a*).

Inclusions of the above types were also observed, which contained a small particle of a metallic phase (Fig. 3*b*).

Ingot SE—The inclusions were mainly spherical and occurred in two forms :

(*a*) Opaque dark-grey inclusions, some of which contained metallic particles (Fig. 3*c*). These inclusions, few in number, were crystalline fayalite, a fact later confirmed by etching tests. Occasionally small inclusions were seen that contained traces of the eutectic structure, which probably consisted of fayalite and the fayalite–wüstite eutectic.
(*b*) Transparent glassy inclusions, sometimes forming in groups. The colour varied from deep orange-brown, through green to a pale green-white. The darker colours were associated with low transparency, whilst the green-white colour was associated with a higher degree of transparency (Fig. 3*d*). A group of these glassy inclusions is shown in Fig. 3*e*.

Ingot SG—All the inclusions were spherical, the smaller ones tending to form groups. Many particles showed the presence of two glass phases within the one inclusion (Figs. 4*a* and 4*b*) : these were probably end members of the two-phase liquid region in the oxide system. Varying degrees of transparency were observed (Fig. 4*c*) and a slight green colour was exhibited by some of the inclusions.

Ingots SL and SM—Both of these ingots contained much smaller randomly distributed spherical inclusions of a very high degree of transparency. They varied from pearl to colourless and showed marked internal reflections (Fig. 4*d*). As confirmed by later tests, these inclusions were highly siliceous. No evidence was found of two glasses occurring within the same particle.

Examination under Polarized Light

The inclusions in ingots *SA* and *SC* were opaque and did not show bright under crossed nicols. In ingot *SE* the dark grey inclusions did not polarize, but the glassy inclusions all polarized under crossed nicols with varying degrees of brightness (Fig. 5*a*). The orange-brown inclusions polarized only faintly with a deep orange colour, whilst the lighter green-white inclusions polarized brightly white. In every case, polarization under crossed nicols resulted in the dark cross and concentric dark circles typical of globular transparent inclusions. The positions of the arms of the cross did not alter when the specimen was rotated.

The inclusions in ingot *SG* showed three main effects under polarized light and crossed nicols :

(i) The monophased glassy inclusions showed the typical dark cross and concentric circles (Fig. 5*b*)
(ii) The presence of the second glass phase within the main inclusion sometimes caused part of the dark cross to be obscured (Fig. 5*c*)
(iii) In other inclusions the second glass phase did not have any effect upon the dark cross.

In both the ingots *SL* and *SM* the inclusions polarized very brightly under crossed nicols with the typical dark-cross effect (Fig. 5*d*). The colour under crossed nicols was always white.

Whiteley's Silver Nitrate Reagent for Sulphides

None of the ingots containing the glassy inclusions showed any reaction to Whiteley's test,[4] and ingots *SA* and *SC* showed the presence of only minor amounts of sulphide.

Alcoholic 10% Nitric Acid Solution

This reagent did not attack any of the inclusions, but they were caused to stand in heightened relief.

Table I

X-RAY ANALYSIS OF RESIDUES FROM INGOTS DEOXIDIZED BY SILICON
From the data of Evans and Sloman[1]

Ingot	Wüstite (FeO)	Magnetite (Fe_3O_4)	Fayalite ($2FeO.SiO_4$)	α-Tridymite (SiO_2)	*x* (Unidentified)	Total Silicon Content, wt.-%
SA	Large	Medium	...	...	...	0·0039
SC	Large	Medium	Medium	...	...	0·017
SE	Medium	Trace	...	Medium	Small	0·034
SG	Trace	...	...	...	...	0·050
SL	...	Large	Medium	Trace	Medium	0·076
SM	...	Medium	...	Medium	Small	0·10

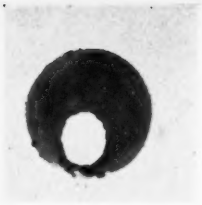
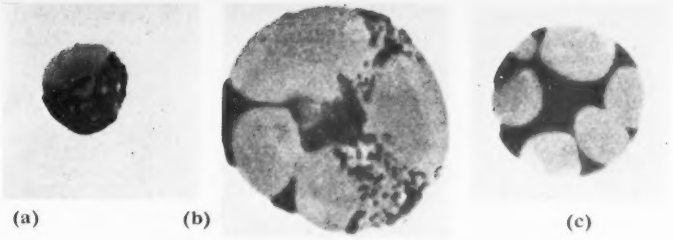
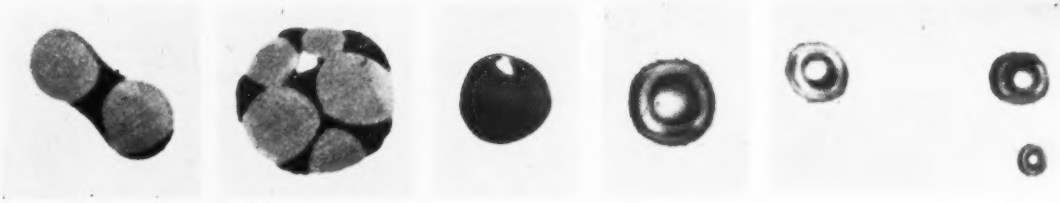


Fig. 1—Ingot *SA*: Wüstite globule containing an inner metallic particle × 1000



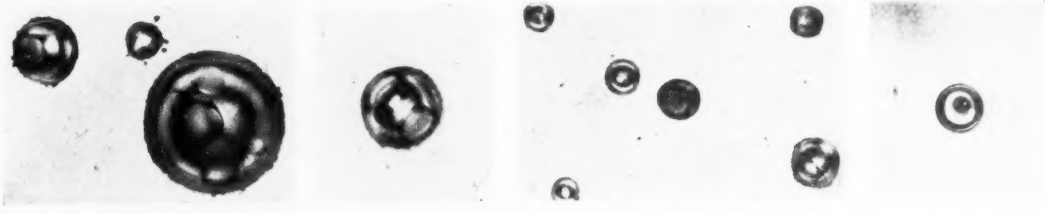
(a) (b) (c)

Fig. 2—Ingot *SC*: (*a*) Duplex globule of wüstite and wüstite–fayalite eutectic; (*b*) duplex inclusion of eutectic and degenerate eutectic; (*c*) duplex inclusion of wüstite and degenerate eutectic; O.I. × 2000



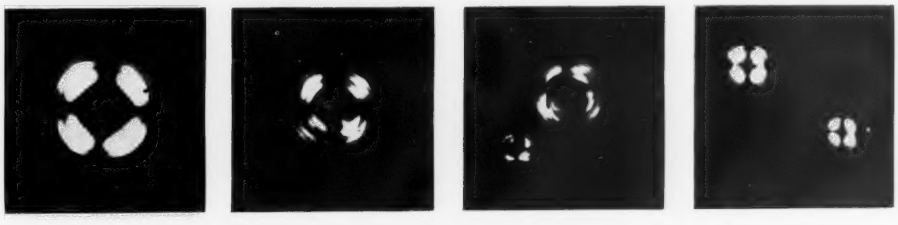
(a) (b) (c) (d) (e)

Fig. 3—Ingot *SC*: (*a*) Dumb-bell shaped duplex inclusion of wüstite and fayalite; (*b*) globular inclusion of primary wüstite and degenerate eutectic, containing metallic particle. Ingot *SE*: (*c*) Crystalline fayalite containing metallic particle; (*d*) spherical translucent glassy silicate; (*e*) degrees of translucency of glassy silicate inclusions; O.I. 2000



(a) (b) (c) (d)

Fig. 4—Ingot *SG*; (*a*) Spherical two-phase translucent inclusions (× 1500); (*b*) smaller two-phase glassy silicate (× 2000); (*c*) glassy inclusions of varying translucency, some two-phase (× 1500); (*d*) Ingot *SM*: Spherical colourless glassy silicate high in silica (× 2000); O.I.



(a) (b) (c) (d)

Fig. 5—(*a*) Ingot *SE*: Glassy silicate (× 2000); Ingot *SG*: (*b*) Normal monophase glassy silicate; (*c*) distortion of dark cross by the two glassy phases (× 1500); (*d*) Ingot *SM*: Highly siliceous inclusion (× 2000); O.I., under crossed nicols

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(*a*) Opaque dark-grey inclusions, some of which contained metallic particles (Fig. 3*c*). These inclusions, few in number, were crystalline fayalite, a fact later confirmed by etching tests. Occasionally small inclusions were seen that contained traces of the eutectic structure, which probably consisted of fayalite and the fayalite–wüstite eutectic.
(*b*) Transparent glassy inclusions, sometimes forming in groups. The colour varied from deep orange-brown, through green to a pale green-white. The darker colours were associated with low transparency, whilst the green-white colour was associated with a higher degree of transparency (Fig. 3*d*). A group of these glassy inclusions is shown in Fig. 3*e*.

Ingot SG—All the inclusions were spherical, the smaller ones tending to form groups. Many particles showed the presence of two glass phases within the one inclusion (Figs. 4*a* and 4*b*) : these were probably end members of the two-phase liquid region in the oxide system. Varying degrees of transparency were observed (Fig. 4*c*) and a slight green colour was exhibited by some of the inclusions.

Ingots SL and SM—Both of these ingots contained much smaller randomly distributed spherical inclusions of a very high degree of transparency. They varied from pearl to colourless and showed marked internal reflections (Fig. 4*d*). As confirmed by later tests, these inclusions were highly siliceous. No evidence was found of two glasses occurring within the same particle.

Examination under Polarized Light

The inclusions in ingots *SA* and *SC* were opaque and did not show bright under crossed nicols. In ingot *SE* the dark grey inclusions did not polarize, but the glassy inclusions all polarized under crossed nicols with varying degrees of brightness (Fig. 5*a*). The orange-brown inclusions polarized only faintly with a deep orange colour, whilst the lighter green-white inclusions polarized brightly white. In every case, polarization under crossed nicols resulted in the dark cross and concentric dark circles typical of globular transparent inclusions. The positions of the arms of the cross did not alter when the specimen was rotated.

The inclusions in ingot *SG* showed three main effects under polarized light and crossed nicols :

(i) The monophased glassy inclusions showed the typical dark cross and concentric circles (Fig. 5*b*)
(ii) The presence of the second glass phase within the main inclusion sometimes caused part of the dark cross to be obscured (Fig. 5*c*)
(iii) In other inclusions the second glass phase did not have any effect upon the dark cross.

In both the ingots *SL* and *SM* the inclusions polarized very brightly under crossed nicols with the typical dark-cross effect (Fig. 5*d*). The colour under crossed nicols was always white.

Whiteley's Silver Nitrate Reagent for Sulphides

None of the ingots containing the glassy inclusions showed any reaction to Whiteley's test,[4] and ingots *SA* and *SC* showed the presence of only minor amounts of sulphide.

Alcoholic 10% Nitric Acid Solution

This reagent did not attack any of the inclusions, but they were caused to stand in heightened relief.

Table I

X-RAY ANALYSIS OF RESIDUES FROM INGOTS DEOXIDIZED BY SILICON
From the data of Evans and Sloman[1]

Ingot	Wüstite (FeO)	Magnetite (Fe_3O_4)	Fayalite ($2FeO.SiO_2$)	α-Tridymite (SiO_2)	*x* (Unidentified)	Total Silicon Content, wt.-%
SA	Large	Medium	...	...	...	0·0039
SC	Large	Medium	Medium	...	...	0·017
SE	Medium	Trace	...	Medium	Small	0·034
SG	Trace	...	...	...	...	0·050
SL	...	Large	Medium	Trace	Medium	0·076
SM	...	Medium	...	Medium	Small	0·10

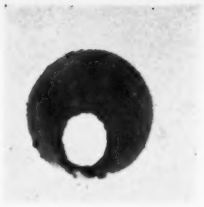
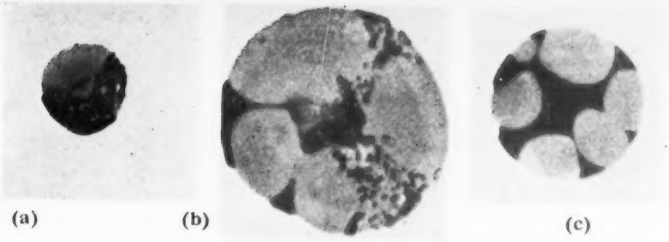


Fig. 1—Ingot *SA*: Wüstite globule containing an inner metallic particle × 1000

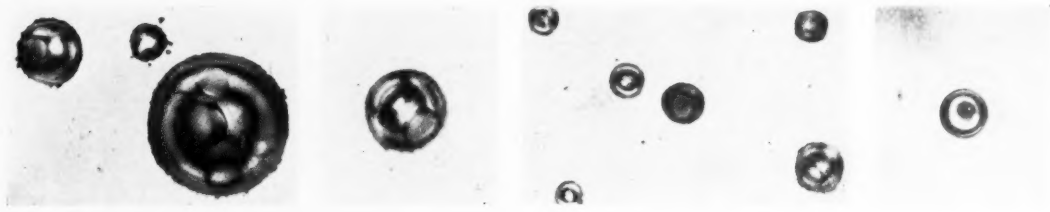


(a) (b) (c)

Fig. 2—Ingot *SC*: (*a*) Duplex globule of wüstite and wüstite-fayalite eutectic; (*b*) duplex inclusion of eutectic and degenerate eutectic; (*c*) duplex inclusion of wüstite and degenerate eutectic; O.I. × 2000

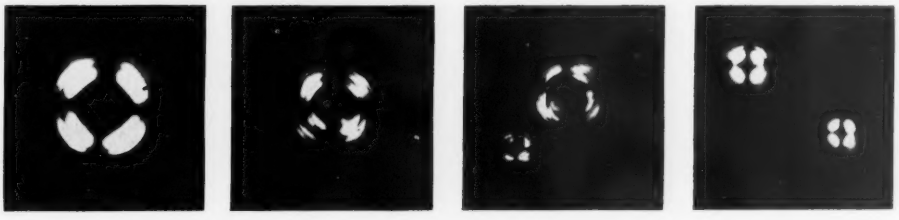
(a) (b) (c) (d) (e)

Fig. 3—Ingot *SC*: (*a*) Dumb-bell shaped duplex inclusion of wüstite and fayalite; (*b*) globular inclusion of primary wüstite and degenerate eutectic, containing metallic particle. Ingot *SE*: (*c*) Crystalline fayalite containing metallic particle; (*d*) spherical translucent glassy silicate; (*e*) degrees of translucency of glassy silicate inclusions; O.I. × 2000



(a) (b) (c) (d)

Fig. 4—Ingot *SG*; (*a*) Spherical two-phase translucent inclusions (× 1500); (*b*) smaller two-phase glassy silicate (× 2000); (*c*) glassy inclusions of varying translucency, some two-phase (× 1500); (*d*) Ingot *SM*: Spherical colourless glassy silicate high in silica (× 2000); O.I.



(a) (b) (c) (d)

Fig. 5—(*a*) Ingot *SE*: Glassy silicate (× 2000); Ingot *SG*: (*b*) Normal monophase glassy silicate; (*c*) distortion of dark cross by the two glassy phases (× 1500); (*d*) Ingot *SM*: Highly siliceous inclusion (× 2000); O.I., under crossed nicols

Table II

IDENTIFICATION OF INCLUSIONS IN IRON–SILICON–OXYGEN ALLOYS

Type of Inclusion	Common Occurrence	Opacity	Colour under Normal Illumination	Appearance under Polarized Light and Crossed Nicols	Action of Silver Nitrate (Whiteley's Test)	Reaction to Etching With		
						10% Alcoholic HNO_3	5% Alcoholic HCl	20% Aqueous HF and 5% NaF
Solid solution rich in wüstite, FeO	(i) Rounded particles containing metallic areas (ii) Rounded particles in duplex wüstite – fayalite inclusions	Opaque	Light to medium grey	Black	No reaction	No reaction but stands in heightened relief	More or less completely attacked	Virtually no reaction
Fayalite, $2FeO.SiO_2$	(i) Eutectic or degenerate eutectic in duplex inclusions with wüstite (ii) Rounded particles of crystalline fayalite	Opaque	Dark grey	Black	No reaction	No reaction but stands in heightened relief	No reaction	Attacked
Silicates of indeterminate composition	Rounded glassy inclusions, showing two-glass phases in certain composition ranges	Varying degrees of transparency with marked internal reflections	Colourless in the high SiO_2 inclusions, through green to dark orange-brown. The darker the inclusion the less the transparency	All polarize brightly with dark cross and concentric dark circles. The colour under crossed nicols varies from white, through green to dark orange. The deeper the colour the less bright the appearance under crossed nicols. The presence of a second glass phase may or may not affect the dark cross and concentric circles	No reaction	No reaction but stands in heightened relief	No reaction	Attacked

This effect was particularly noticeable in the wüstite inclusions in ingot *SA*.

Alcoholic 5% Hydrochloric Acid Solution

The wüstite inclusions in ingot *SA* were completely dissolved. In ingot *SC* the wüstite of the duplex inclusions could be preferentially attacked provided that a carefully controlled etching procedure was used. Over-etching, however, rapidly led to a complete removal of the inclusion. Inclusions consisting of wüstite and the fayalite–wüstite eutectic were completely attacked, since as the wüstite of the eutectic was soluble, the whole of the eutectic was removed. On the other hand, inclusions consisting of wüstite and the degenerate eutectic only showed attack on the wüstite, the fayalite remaining unattacked. The careful control of the etching conditions must, however, be emphasized.

The glasses in ingots *SE*, *SG*, *SL*, and *SM*, and the fayalite in ingot *SE*, were unattacked by this solution, but in ingot *SG* the interface between the two glass phases seemed to be more clearly defined. This may have been due to the slight attack occurring where the interface between the two phases cut the surface of the inclusion.

Aqueous 20% Hydrofluoric Acid Solution

The glass inclusions in ingots *SG*, *SL*, and *SM*, together with the crystalline fayalite and glasses in ingot *SE*, were all completely dissolved by the hydrofluoric acid. In ingot *SC* the eutectic and degenerate eutectic portions of the inclusions were preferentially attacked to the wüstite provided that very careful control of the etching was again maintained. Over-etching resulted in the complete removal of the whole inclusion.

Aqueous 5% Sodium Fluoride Solution

The attack by this solution on the glasses was of a nature similar to, but slower than, that of hydrofluoric acid. Selective etching of the two glasses was not achieved.

Aqueous Alkaline Solutions

The reagents used were 10% sodium hydroxide solution and alkaline sodium picrate. Neither, used in cold solution, attacked the inclusions, and attempts to promote chemical attack by electrolysis were unsuccessful, resulting only in pitting of the base metal.

A summary of these experimental results is given in Table II.

DISCUSSION OF THE TYPES OF INCLUSIONS PRESENT

The whole range of inclusions represented by the deoxidation of iron by silicon has been found in the specimens examined. With increasing silicon content the inclusions varied from single-phase wüstite, through duplex fayalite–wüstite to fayalite, and subsequently to inclusions which were glassy. Within certain ranges of silica content, two glass phases were observed to co-exist within one inclusion. These corresponded to two liquid phases in the $FeO–SiO_2$ system.

The magnetite which was reported by X-ray analysis to be present in all ingots except *SG* was not observed microscopically and, as in the study of aluminous inclusions,[3] its mode of occurrence remains unknown.

Ingot *SA*

The inclusions were mainly wüstite (FeO), often containing an inner metallic particle. The possible reasons for the occurrence of these metallic particles have been discussed previously.[2]

Ingot *SC*

The inclusions were mainly duplex in nature, consisting of wüstite globules in a matrix of eutectic or degenerate eutectic of fayalite–wüstite. Occasionally these inclusions contained an inner metallic particle. An interesting feature of the inclusions was the protuberance of the fayalite beyond the perimeter of the wüstite, an effect which may have been caused by surface tension.

Ingot *SE*

This ingot contained inclusions which were mainly either fayalite or of a glassy nature. The fayalite did not polarize and sometimes contained small metallic particles. The great variation in colour and transparency of the inclusions may have been due to variations in their thickness and silica content. Under crossed nicols the glasses showed the typical dark cross and concentric circles with colours varying from orange, through green to white. These colours were probably solely the transmission colours of the glass and not a true anisotropic phenomenon.

Ingot *SG*

Many of the two-phase glasses were observed in this ingot, associated with the two-liquid region in the $FeO–SiO_2$ system. The refractive indices of the two phases were sufficiently different to reveal the interface between them, but under crossed nicols the second phase exerted an indeterminate effect upon the dark cross and concentric circles. Similar colour and transparency effects were observed to those described for ingot *SE*, although the inclusions were generally lighter coloured, probably because of their greater transparency and associated internal reflections.

Ingots *SL* and *SM*

The inclusions were quite small and highly transparent, polarizing white under crossed nicols with dark cross and concentric dark rings.

Acknowledgments

The authors wish to thank Mr. F. H. Saniter, Director of Research, The United Steel Companies Ltd., for permission to publish this paper ; they also acknowledge the assistance received from the members of the Solid Steel Study Group of the former Gases and Non-Metallics Sub-Committee.

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1. E. Ll. Evans and H. A. Sloman : *J. Iron Steel Inst.*, 1952, vol. 172, pp. 296–300.
2. R. E. Lismer and F. B. Pickering : *Ibid.*, 1952, vol. 170, pp. 48–50.
3. R. E. Lismer and F. B. Pickering : *Ibid.*, 1952, vol. 172, pp. 381–383.
4. J. H. Whiteley : *Ibid.*, 1948, vol. 160, pp. 365–366.

DISCUSSION ON PAPERS

DISCUSSION AT MEETINGS WRITTEN CONTRIBUTIONS

Special Meeting in Sweden, 1954

A SPECIAL MEETING OF THE IRON AND STEEL INSTITUTE was held in Sweden from Monday, 7th June, to Friday, 18th June, 1954, by invitation of the President and Council of Jernkontoret. A Joint Technical Session was held at Riksdagshuset, Stockholm, on Saturday, 12th June, at which papers by British and Swedish authors were presented and discussed. The first group of papers were on the pelletizing of iron-ore concentrates and the second on low-alloy steels. The discussion on the latter group is printed below; that on the pelletizing papers will be printed in a subsequent issue of the *Journal*.

Joint Discussion on the Papers—

TENSILE AND IMPACT PROPERTIES OF Fe–Si, Fe–Ni, Fe–Cr, AND Fe–Mo ALLOYS OF HIGH PURITY*

By W. P. Rees, B. E. Hopkins, and H. R. Tipler

EFFECT OF HEAT-TREATMENT ON THE BRITTLENESS OF HIGH-PURITY IRON–NITROGEN ALLOYS †

By B. E. Hopkins and H. R. Tipler

PEARLITE-FREE BASIC BESSEMER STEEL: ITS FABRICATION AND PROPERTIES‡

By Å. Josefsson

Mr. **B. E. Hopkins** (National Physical Laboratory) introduced the first two papers and Mr. **Å. Josefsson** (Stora Kopparbergs Bergslags AB, Sweden) the third.

Mr. **D. A. Oliver** (B.S.A. Group Research Centre): These three papers deal with very pure (or nearly pure) systems, and it is interesting that physical metallurgy, having particular reference to fundamentals, is calling for information of the kind presented, so that in this connection each paper makes an important and welcome contribution. Many years ago Dr. Smithells wrote his famous book on the effects of impurities in metals, which called attention to the major influence on physical properties of very small quantities of impurities. More recently, Dr. N. P. Allen and his colleagues at the National Physical Laboratory have done excellent work, supported by the British Iron and Steel Research Association, and the papers presented by Mr. Hopkins form part of that unfolding story.

A short time ago, Mr. Jerome Strauss, who gave the Second H. W. Gillett Memorial Lecture, coined a term for this general subject, namely, 'micro-metallurgy' or 'the metallurgy of minute additions.' We might equally consider 'micro-metallurgy' to be 'the metallurgy of minute subtractions,' because great care and effort need to accompany researches of this character, in order to confer fundamental significance on them and to ensure that the deductions drawn from them are valid.

Mr. Josefsson's admirable and comprehensive paper is a pleasure to read. His research starts in the laboratory and ends in the pilot plant, and his account of it holds an excellent balance between academic and practical considerations.

The broad concept underlying his paper is that we can add what are generally regarded as impurities (in this case phosphorus and nitrogen) and secure useful strengthening properties from them, provided that an unusually low carbon limit is also maintained. It is true that some impurities confer increased yield strength or tensile strength, but I searched for even passing reference to the mechanical properties secured in the transverse direction to that of (hot) forging or rolling. It may be that in constructional steels these properties are not of primary concern, but when examining a new type of steel an indication of the transverse properties

* *J. Iron Steel Inst.*, 1954, vol. 177, May, pp. 93–110.
† *Ibid.*, pp. 110–117.
‡ *Ibid.*, pp. 118–128.

in relation to those in the longitudinal direction would have been most valuable. Can the author give any indication of how the mechanical properties in the two directions actually compare ?

A further point of particular interest to me was the selected carbon limit of 0·015%. From the paper, I could not decide whether this was deduced from high- or low-temperature considerations, but from some work I undertook myself some years ago I arrived at a carbon figure of 0·007%, at which the first signs of cementite formation occurred. My own observations were based on magnetic measurements, which indicated that below 0·007% the carbon atoms stayed substantially in random solid solution. Mr. Josefsson's figure is almost double the one I found, but it may well be that his figure is correct for all practical purposes, although the lower figure may be valid on strictly academic considerations. Would he care to comment on this point ?

On p. 120, the use of titanium is mentioned as a means of combating the effects of carbon when the carbon is on the high side. In steel *B* (Table II) there is recorded a titanium content of 0·27%, which is sufficient not only to combine with all the carbon, but also to leave some free titanium. If free titanium is present, it would combine strongly with any nitrogen in the system, so that the titanium addition may also have neutralized the nitrogen present and converted it into titanium nitride. The addition of titanium may not invalidate the objective of controlling the carbon, but the dual role should, I submit, be kept in mind.

On p. 127, reference is made in the welding tests to electrode type KB 5225 (Basic). Is this of the type usually called ' hydrogen-controlled ' or ' low-hydrogen ferritic ' with a lime-rich coating ?

The more practical aspects of this research in relation to Bessemer steelmaking will be left to those better qualified to comment.

The two papers by the N.P.L. workers record work which is factual in character, and which sets a very high experimental standard. In Fig. 18 (p. 108) the effects of alloying additions in pure iron are presented, and the fact that the yield and the tensile strength values follow one another is, to me, very significant. This is not likely to be true if appreciable impurities were present, and possibly that is why the curves, at first sight, seem so unusual. For instance, the initial negative effect of chromium additions would not be expected to occur if appreciable traces of carbon had been left behind in the samples. The interesting point now arises as to what these carefully observed curves really mean; have the authors considered the effect of micro-segregation in relation to a particular added element ? If one element is added the atoms may be distributed uniformly throughout the grains; if another is added there may be a tendency for it to be precipitated at, or concentrated round, the grain boundaries, or even to form dense zones or banded structures. The work might be enhanced in value if a limited amount of X-ray micro-radiography could be undertaken on selected samples. In the past, whenever I have investigated samples of a similar nature, I have invariably found that micro-segregation is playing a part. Molybdenum and manganese, for example, are elements which segregate on a micro-scale in a very surprising way. It seems almost impossible to add molybdenum to some ferrous alloy systems without obtaining concentrated zones or bands of at least three-fold concentration. Chromium, on the other hand, seems to prefer to migrate to grain boundaries, whilst silicon would appear to distribute itself much more uniformly. It would be helpful to have the authors' views on this aspect.

In regard to the paper by Hopkins and Tipler, it does seem important to decide when the nitrogen atoms should be regarded as in solid solution and uniformly distributed through the system, or when they should be regarded as having formed stable chemical compounds, in which case we are then concerned with the remaining nitrogen atoms which can be regarded as still free. A possible intermediate stage is where nitrogen atoms combine to form molecules in fissures, where there is a possible openness of structure, such as at grain boundaries; then nitrogen gas pressure might well be induced in these elementary cavities. If this last state of affairs arises, we have also to consider the effect of internal stresses, induced by the nitrogen gas pressures, on the mechanical processes. The curious intermediate phase of brittleness might be explained by this last mechanism, because lowering the temperature might, on balance, reduce any cavity gas pressure besides precipitating iron nitrides, whilst raising the temperature might well increase the solid solubility and also promote dissociation in nitrogen molecules. Do the authors think this suggestion helps to explain their observations ?

On p. 117, the authors state: " The simplest explanation of the results at present is that there is a maximum tendency for nitrogen to diffuse to the grain boundaries at 600–700° C., whereas above and below this temperature range the tendency to diffuse away from the grain boundaries becomes progressively stronger." It is easy to accept this statement for increasing temperatures, but is it probable for *decreasing* ones ? Further comment on this point from the authors would be valuable, as least energy considerations seem to indicate the reverse tendency.

Professor **E. Rudberg** (Metallografiska Institutet, Stockholm, Sweden): The research group at the N.P.L. is engaged on an extensive programme to elucidate the effect upon mechanical properties of increasing additions of a number of alloying elements to very pure iron. Their two papers constitute the most recent contributions according to this programme. We in Sweden consider this programme to be very important; we greatly appreciate the care and skill with which these difficult problems are attacked. The value of these new contributions extends beyond the practical field; they contain much information which should be studied with interest by those concerned with the theory of the strength of materials.

As conscientious experimenters the authors are careful to present their results without the addition of any detailed theory, which, at the present stage, might unintentionally obscure some of the facts. But, for the purposes of discussion, the results must be arranged in some kind of system, involving the use of a model. Mentally, the loaded test piece is regarded as a chain containing links of three types. When subjected to the proper load the first type will fail by ' ductile ' cone-and-cup fracture, the second type by ' brittle ' cleavage through the grains, and the third by intergranular fracture. The authors exercise both care and ingenuity in deriving from the behaviour of test bars at different temperatures the strength of the third link, the grain-boundary strength. In using such a model it is well to remember that the interplay of the three fracture processes envisaged may well hold elements other than those of the series loading. Thus comparatively small initial plastic deformations could conceivably condition the original grain-boundary region for the intercrystalline failure eventually observed.

In the first paper, Fig. 17 gives a comparison of (*a*) 0·5% proof stress and (*b*) tensile strength, as a function of the amount of addition of Cr, Mo, Ni, Mn, and Si. All these curves start from the values for pure iron

(concentration 0). The data were taken at room temperature, except in the case of Si, where the temperature was 200° C. Yet all the curves in both Fig. 17*a* and *b* start from the same point, at concentration 0. Does this mean that the corresponding values for pure iron are entirely unaltered in going from room temperature (say, 20° C.) to 200° C. ?

In Fig. 12 of the first paper several ferrite grains are visible, the boundaries of adjoining grains showing up clearly. There are also some weaker contours, which are said to mark the boundaries of the austenite grains, now no longer in existence, out of which the ferrite grains developed by transformation. Has any relation been observed between the two kinds of boundary ?

This would seem to be of interest in connection with the experiment reported in Fig. 3 of the second paper. In selecting values of equivalent holding time for nitrogen diffusion in the test pieces at a series of different temperatures (*i.e.*, time inversely proportional to the diffusion coefficient), the authors have in general decreased the time with increasing temperature. For 1000° C., however, the time has a value more than five times that for 900° C. This choice would seem to be based on the consideration that diffusion at the higher temperature no longer takes place in ferrite, but in austenite, where diffusion is more difficult. The observed low fracture strength in Fig. 3 is interpreted as being due to the grain boundaries having become weakened by diffusion of impurity (presumably nitrogen) to the grain boundaries, which at the temperatures of test must be ferrite boundaries. At 1000° C. the boundaries present must be austenite boundaries. Unless the final location of the ferrite boundaries (not yet in existence but soon to come, when the temperature is lowered) makes itself felt in advance, it is difficult to see how the diffusing nitrogen at 1000° C. knows where it is to go. Information on any observed or expected relation between austenite grain boundaries and later ferrite boundaries would be welcome on this point.

From the observed speed with which the 600° C. brittleness establishes itself, the authors of the second paper have computed the depth of the lamellar region next to the boundary from which impurities have time to reach the boundary, given reasonable figures for the diffusion coefficients. The result of this—that the depth must be rather small—seems quite convincing. The experimentally determined time for brittleness to develop in the tests illustrated in Fig. 3 is definitely too short to make it at all likely that diffusion of any of the common elements from the interior of large grains to the grain boundaries is responsible for this embrittlement. In fact, these experimental results appear to rule out any explanation along such lines. In view of this, the statement about the likely cause of the observed effect of grain size, given on p. 115, seems surprising, unless the diffusing constituent here is something quite different from that which produces the 600° C. brittleness. Are there no other suggestions than these chemical ones for the possible cause of the grain-boundary weakness in large-grained samples ?

Mr. W. E. Bardgett (United Steel Companies Ltd.): On reading the paper by Rees, Hopkins, and Tipler many questions immediately present themselves, amongst which the most interesting is that of brittleness in alloys being accompanied by intergranular failure. It appears that many metallurgical phenomena can give rise to intergranular failure, prominent amongst which are temper brittleness, overheating and burning, and intergranular embrittlement in steel castings due to the formation of AlN.

It would therefore appear to be essential to qualify any change in mode of fracture from ductile to brittle by stating whether the brittle fracture is cleavage (transgranular) or intergranular. Much of the present work is concerned with intergranular fracture, sometimes in large proportions, and it is therefore difficult to fit this work into the general picture of brittle fracture in mild-steel ship plate (which I understand is cleavage), as was probably intended.

Considering the two modes of brittle fracture (cleavage and intergranular), what are the authors' views about the possibility of a second transition temperature from one type of brittle fracture to the other ? They state that intergranular fracture is accompanied by:

(i) A scatter of the results

(ii) Not so sharp a transition temperature.

Have they examined the lower limb of the transition curves in detail with a view to detecting any sign of a second transition from intergranular to cleavage, or *vice versa*, especially in the silicon alloys ? We have found double inflexions in transition curves of low-carbon molybdenum–boron steels, which might indicate such a second transition, and it is proposed to look into the matter further to try to detect such a change.

The effect of increased grain size is most marked on the transition temperature and, as it is well known that a smaller area of grain boundary aids intergranular fracture, it would be of interest to hear the authors' views on whether, even in normal commercial steels, the effect of increasing grain size is merely to change the brittle fracture from cleavage to intergranular, with a consequent rise in the transition temperature.

I should have expected that in the silicon alloys used in the investigation there was enough carbon to give a yield point, but there appears to be a contradiction between the text and Table II, in which yield points are quoted. In Fig. 1 it appears that the effect of temperature on the tensile strength alters sign at 2–3% of silicon; can the authors give the reason for this, and also for the apparent anomalous ductility values of the 5% silicon alloy at high temperatures ?

The authors state that the silicon alloy strip cold-rolled more readily in the form of thin sheet; might this be due to preferred orientation ?

In general, the strength and ductility of the alloys indicate solid-solution hardening, but is there a rational explanation for the reverse effect of silicon, compared with chromium and molybdenum, on the transition temperature, although all three elements close the γ loop. However, silicon is a smaller atom than iron, whilst chromium and molybdenum are larger, and this might possibly explain the different abilities of the lattices in the neighbourhood of the grain boundaries to adsorb impurities.

The method of producing a fine grain size in the silicon alloys was to cold-work and then anneal at 700° C.; this method, by causing the grain boundaries to alter position at a relatively low temperature, may not have been conducive to diffusion of impurities to the new grain boundaries, and may therefore have resulted in cleavage failure. Did the authors anneal for longer periods at 700° C., when, if the quantity of impurities was sufficient, diffusion to the grain boundaries would have occurred and presumably intergranular fracture would have been the result ?

In regard to the nickel alloys, was it really possible to measure the grain size of the acicular α_2 structure, or was the grain size reported more likely to be the prior γ grain size. Also, is a 'star' fracture typical of the α_2 in both the manganese and nickel alloys ? I agree that tempering would tend to remove some of the lattice distortion in the α_2, but might it not also lead to an

embrittling action similar to that described for the nitrogen-containing alloys of the second paper ?

The change in brittle fracture from cleavage to intergranular at 2% of nickel is strange and may have been due to the α_2 structure. The effect of nickel in commercial steels is invariably to decrease the transition temperature, but Fig. 9*a* shows that in these pure iron alloys the effect of nickel is the opposite. Can the authors throw any light upon this apparent anomaly ? If the effect is due to the formation of the α_2 structure, are low-carbon acicular-type structures always liable to give poor impact properties ?

It seems strange that in the molybdenum alloys the allotropic change does not have some effect on the mechanical properties, due to some transformation giving a change in grain structure. Further to this point of the allotropic changes, the fact that some alloys (*e.g.*, nickel and manganese) will show a change point in the higher-percentage alloys, whilst the molybdenum alloys probably will not, does not lead to ease of interpretation of the results for all the alloys as a whole.

Finally, with regard to the effect of solid-solution strengthening, the figures for atomic diameters (taken from Hume-Rothery[1]) are as follows:

Iron	2·52–2·48 Å.
Silicon	2·34 Å.
Nickel	2·49 Å.
Chromium	2·57 Å.
Manganese	2·75 Å.
Molybdenum	2·80 Å.

Figure 18 of the paper places the alloying elements in the following order, as far as their strengthening effect is concerned: Mo, Mn, Si/Ni, Cr. From the sizes of the atoms, the difference of which from iron might be expected to give a measure of the solid-solution hardening, the following order was found: Mo, Mn, Si, Cr, Ni. It appears, therefore, that for obviously large atoms the size factor is quite important; whilst this agrees with the authors' postulation that lattice distortion may not be the main strengthening effect, it does seem that in these solid solutions lattice distortion has a considerable strengthening effect.

Probably the most important phenomenon discussed in the paper by Hopkins and Tipler is the decrease in intergranular fracture strength after reheating at temperatures in the vicinity of 600° C., to which I shall refer later.

Do the authors think that the higher yield stress in the furnace-cooled condition, compared with the air-cooled condition, is solely due to precipitation hardening ?

In general, the impact transition temperature is lowered as more and more precipitation is allowed to occur. Besides the effect of grain size, as mentioned by the authors, I should like to hear their views as to whether this phenomenon (lowering of the transition temperature) can be either:

(i) Due to the cleansing of the lattice by the removal of nitrogen as precipitate (say, $Fe_{16}N_2$), thereby lessening the probability of diffusion of nitrogen to the grain boundaries; or
(ii) Simply a result of the precipitation hardening.

The comparison of the effect of rate of cooling on nitrogen alloys on the one hand and carbon and carbon–manganese alloys on the other was interesting, as the former gave intergranular and the latter cleavage fractures. Can the authors explain this difference between the effects of carbon and nitrogen, as both are

(i) Interstitial elements
(ii) Austenite formers
(iii) Precipitation hardeners in small proportions ?

Rapid cooling from 950° C. resulted in a high transition temperature and also presumably in a fairly low fracture stress. It could, of course, be due to a high yield stress resulting from finer grain size or nitrogen in solid solution. This low fracture stress most probably results from a condition present at high temperature and which has been retained by water-quenching. Should it therefore be inferred that at high temperatures of heating the fracture stress again falls ?

The authors postulate that :

(i) Above 700° C. nitrogen is more uniformly in solid solution, owing to thermal energy
(ii) In the brittle range the nitrogen atoms segregate somewhat to the boundaries
(iii) Below the brittle range the nitrogen again moves into the grains more uniformly.

The last seems to be the only one capable of explaining the facts, but I would like to add a little more for their consideration. The nitrogen will have different energy states depending upon:

(i) The reheating temperature
(ii) Whether it is in solid solution in an interstice, at a grain boundary, or around a dislocation.

It is postulated, therefore, that in the brittle temperature range, the nitrogen atoms at grain boundaries are most stable (*i.e.*, have least energy), but below this range the more stable position of a nitrogen atom is at a dislocation within the grain. Thus below the brittle range the nitrogen tends to move away from the grain boundaries. This does not apparently offend the authors' conclusion that such clustering, if it is occurring, may not necessarily be an embryo nucleus for $Fe_{16}N_2$ formation. Furthermore, have the authors considered the possibility that the nuclei of $Fe_{16}N_2$ may arise from a type of ordering of the nitrogen atoms around the dislocations, which have such an orientation that the observed orientation effects of $Fe_{16}N_2$ in ferrite would be observed ?

I agree entirely with the authors that 100% adsorption to the grain boundaries is not necessary to obtain a low fracture stress, and that diffusion of nitrogen would only be necessary over a small distance, thereby accounting for the rapidity of the embrittling reaction

Monsieur **C. Crussard** (IRSID, France): The paper by Josefsson is one of the most interesting contributions to steel problems I have ever read, and is of outstanding practical importance. I shall, however, concentrate my discussion on its fundamental implications.

My first point concerns the fine 'submicroscopic' carbide precipitation, which is supposed to give a low transition temperature. It is necessary to be cautious about arguments such as those developed on p. 120 of the paper, in order to prove that a low transition temperature is due to some special kind of dispersion of carbides, for the following reasons. The brittle fracture of a Charpy specimen requires one set of conditions for initiation of the crack, and another set for its propagation. Both sets generally do not coincide. A study of this problem now in progress at IRSID suggests that an abrupt drop (total or partial) in the impact-energy curve means that, at that particular temperature, something is changing in the initiation mechanism. But if two different heat-treatments give the same temperature for this drop (in the present case, treatment for boundary cementite and complete decarburization) it is impossible to deduce from this fact that the mechanism of fracture is the same in both cases without knowing how the propagation conditions do compare.

Thus, it cannot be said that these two heat-treatments give an equivalent state, and that, starting from this

state, a fine submicroscopic dispersion of carbides produces a shift of the transition temperature to lower values.

The hypothesis about these carbides does not seem entirely proved. Nevertheless, assuming that it is true, let us consider whether the effect of such a dispersion on the trend to brittle fracture could be explained. An explanation is quite simple; we at IRSID believe that two important mechanisms of crack initiation are local ductile fracture (which is an intracrystalline mechanism) and intergranular decohesion (under stress concentrations at the end of a slip band). Both dangers are reduced if a fine dispersed phase blocks the dislocations inside the grains and prevents them from pressing too hard on the boundaries, provided that the precipitation hardening effect is not too great (which would promote another type of initiation, by cleavage).

It is surprising that carbide should precipitate during a rapid quench from the α state. Such a precipitation is not observed on fine wires quenched and tested for internal friction. Nevertheless, in large samples, it is possible that the quenching strains may produce a large number of lattice vacancies, which accelerate greatly the rate of precipitation. This effect is well known in Al–Cu alloys.

Another interesting and paradoxical property of these pearlite-free steels is the fact that, despite the presence of large quantities of nitrogen and phosphorus, they do not strain-age very much. One can even say that, for these steels, the higher the nitrogen content, the smaller the strain ageing, as proved by steel $K104$ (Fig. 1) having a smaller rise in transition temperature than steel $K26$. It may be questioned whether, for these steels, the presence of nitrogen is not beneficial against strain ageing, which would explain the influence of manganese (p. 122 of the paper), as manganese is known to have an affinity for nitrogen in solid solution, and would slow down its diffusion and precipitation.

This surprising effect seems difficult to understand in terms of the normal 'Cottrell clouds' theory. There must be some intergranular effect, the strain-ageing properties of nitrogen being able to develop only if there is already some intergranular precipitation or segregation of carbon. This is hard to believe, but is surely of the greatest importance.

CORRESPONDENCE

Mr. **Å. Josefsson** and Mr. **G. Lagerberg** (Stora Kopparbergs Bergslags AB, Sweden) wrote: The extremely interesting findings of Hopkins and Tipler in their work on iron–nitrogen alloys call attention to the grain-boundary structure as compared with the grain-interior structure. There are two points in the discussion given in this paper upon which we should like to comment.

Firstly, the authors have suggested that the nitrogen atoms gather in the grain boundaries at high temperatures and are fixed there by quenching, whereas slow cooling or ageing at room temperature allow these atoms to diffuse away from the boundaries. The contrary seems to be more likely, for the following reasons.

A grain boundary is generally regarded as a transition region, where the lattice is greatly strained, and may then be equivalent to a zone of very high dislocation density in comparison with the interior of the grain. Even if no second phase can be detected in the boundary, an enrichment of interstitial atoms is therefore probable, as Cottrell atmospheres around these dislocations. Such atmospheres are said to be condensed if all interstices in the line of maximum binding energy along the dislocation are filled with interstitial atoms. According to Cottrell[2] this is the case only up to a certain temperature. For carbon, Mott[3] had estimated this temperature to be about 400° C., and for nitrogen it is probably not higher. Above this temperature, the atmosphere becomes more and more diluted, according to the equation

$$C = C_0 \times e^{-V/kT} \quad \ldots\ldots\ldots\ldots\ldots\ldots (1)$$

where C = Concentration in the dislocation region
C_0 = Average concentration in solid solution
V = Interaction energy between the dislocation and the solute atoms
T = Absolute temperature
k = Boltzmann's constant.

Equation (1), strictly valid only for dilute solutions,[2] shows that the atmospheres are successively dissolved and C is approaching C_0 when temperature is raised.

The higher dislocation density in the grain boundaries will cause a long-range diffusion of nitrogen from the boundaries inwards when the Fe–N alloy is heated to say 500° C. Quenching from this temperature will preclude restoration of the higher nitrogen concentration in the grain boundaries that corresponds to low-temperature equilibrium; to fill the dislocations in the grain-boundary zones to the same extent as those in the grain interior, nitrogen atoms must have time for long-range diffusion, which may be provided by slow cooling or by ageing.

This picture seems also to fit with the statement that the brittle-fracture stress is lower after quenching from, say, 500° C. than after furnace cooling. If the higher nitrogen concentration in the grain boundaries is present as atmospheres, there will be a partial relaxation of the lattice strains there. Consequently, at a boundary with high nitrogen concentration, a greater strain contribution is demanded from the external stress in order to surmount the cohesion forces. Thus the brittle-fracture stress (if this notion is temporarily accepted) increases by slow cooling.

Secondly, the authors have considered the concept of brittle fracture to be governed only by the two magnitudes brittle-fracture stress and yield stress, the former obviously being identical with the 'technical' cohesive strength. This way of thinking seems to be well accepted in the literature and is certainly a valuable aid in understanding the brittleness problem, although it sometimes is questioned as being an over-simplification. In the present case it also leads to certain contradictions. Particularly, it is hard to understand why the brittle-fracture strength could decrease as it does when temperature is lowered (*see* Fig. 4). The reverse would rather be expected, as the force needed to separate two layers of atoms has to be greater the less this force is supported by thermal fluctuations. Moreover, the evaluation of brittle-fracture strength at the impact transition temperatures seems to be rather uncertain. As stated by the authors, one has to rely upon extrapolation from low-temperature data. From this it would be very hard to forecast the order of the transition temperatures in impact. In fact, taking the specimens furnace-cooled from 950° C. to room temperature on one hand, and those where the furnace-cooling was interrupted at 300° C. on the other, a comparison of the proof-stress figures at room temperature and the fracture-stress figures at −196° C. will suggest an order of the transition temperatures which is the reverse of that which has been found experimentally.

In a criticism of the theory of brittle fracture as governed by a simple cohesion strength, Zener[4] has pointed out that fracture is presumably almost always preceded by some plastic deformation, even if this may be rather small and may perhaps not show up in elongation figures. Zener suggests that the stress measured

even at a seemingly brittle fracture is a *flow* stress, *i.e.*, the stress required to give a certain amount of plastic deformation, sufficient to cause the critical degree of crack formation (or what Zener terms as ' crystal damage ').

It looks rather attractive to apply these principles to the present problem. It may then be assumed that intercrystalline fracture in the iron–nitrogen alloys is caused by localized flow in quite narrow zones along the boundaries. Such a localized flow would occur if yielding could start and propagate more easily in the grain-boundary zones than in the grain interior; this difference in flow stress must obviously be assumed to increase when the strain rate increases or the temperature is lowered (since tensile testing at room temperature does not cause brittleness). If the rise in yield stress developed by high strain rates and low temperatures in mild steel is regarded as an effect of solute atoms (carbon or nitrogen) forming Cottrell clouds around the dislocations,* this postulation would imply that in the embrittled specimens of the Fe–N alloys the grain-boundary zones are less anchored by such Cottrell clouds than the grain interior. This condition is just what is to be expected in the quenched specimens from the mechanism outlined above, as the dislocations in the grain-boundary zone become less saturated with nitrogen atoms than those in the grain interior, whilst the difference may be levelled out in the furnace-cooled and in the quenched and aged specimens.

If quenching is done from temperatures above 700° C., there is probably some formation of a nitrogen-rich austenite in the grain boundaries (*cf.* the results of Tsou, Nutting, and Menter[5]). This may counteract the effect of the quenching and account for the slightly increased brittle strength at higher quenching temperatures, according to Fig. 3.

The intergranular embrittlement of quenched Fe–C alloys with less than 0·006% C[6] may probably be explained along the same lines. According to equation (1) the density of the atmospheres at a given temperature will be proportional to the average concentration. The concentration in the grain-boundary zones after quenching will therefore be higher the higher the carbon content, even if the density of the dislocation atmospheres will still be less in the boundary zones than in the grain interior. The effect of differences in atmosphere density upon slip resistance will probably be much greater at low than at high densities; *i.e.*, at higher carbon contents the difference in straining behaviour after quenching between the grain-boundary zone and the grain interior will diminish, which would explain the disappearance of the intercrystalline brittleness at higher percentages of carbon. Reference may be made to Wert,[7] who showed that an increase of carbon in Armco iron from 0·0000% to 0·0005% caused the yield point to increase about 100%, whereas the further step from 0·0005% to 0·0010% corresponded to an increment in the yield point of only about 10%. In this case, only the carbon in solid solution was measured, but according to equation (1) there would have been a proportional amount present as dislocation atmospheres; it is probably the latter which is the agent with respect to slip behaviour.

At present, experimental data are admittedly too few to justify more definite statements about the mechanism. These suggestions are put forward in the hope of further experiments being carried out to prove or disprove their validity.

* The grain-boundary surface may consist of irregularities which are not necessarily to be regarded as pure edge (or screw) type dislocations; nevertheless, their effect upon interstitial atoms will probably be very much the same, and equation (1) will be valid in principle, even if the interaction energy V has other values. On quenching from a temperature at which the nitrogen content is evenly distributed throughout the structure, the greater nitrogen-absorbing capacity of these boundary ' dislocations ' will also deprive the dislocations of ordinary screw or edge type in a narrow zone along the boundary of some of the nitrogen atoms that they would have gathered around themselves in the furnace-cooled condition. Thus, slip in the grain-boundary zones will be favoured by quenching, even if the ' dislocations ' in the true boundary surface are not assumed to start slip.

AUTHORS' WRITTEN REPLIES

Mr. **Hopkins** wrote: The possibility of micro-segregation of the added element discussed by Mr. Oliver in connection with the first paper has been considered and was specifically mentioned in the section on Fe–Si alloys. It appears at present to be more profitable to consider segregation of atoms of small size rather than of nickel, for example, to explain the effects obtained.

It is agreed that the model of series loading as put forward by Professor Rudberg may be affected by such factors as small amounts of plastic deformation, and this has to be kept in mind when interpreting experimental results. His suggestion that a small amount of plastic deformation might condition the material in favour of a grain-boundary failure seems to be a subtle effect and it would be desirable to have a clearer picture of what he has in mind. We have taken some licence in using the yield point and tensile strength for the Fe–Si alloys at 200° C. with a zero for iron at room temperature. This is a minor point, however, in that it could only affect the bottom end of the curve for the effect of silicon in Fig. 17, and any small differences between the values for iron at room temperature and 200° C. would not alter the general trend.

Mr. Bardgett finds it difficult to fit the work reported in the first paper into the general picture of brittle fracture in mild-steel ship plate because he assumes that brittle fractures in ship plate are of the cleavage type whereas partly intergranular failures were obtained in this work. We consider that some attention should be paid to the extent to which weakness of grain boundaries affects the brittle fracture of ship plate. It may be significant that although ship plate only rarely comes within our sphere of work, we have drawn attention to two examples of brittle fracture in ship plate taking place partly along the grain boundaries.[8, 9] Short[10] and Borione, Morillon, and Plateau[11] have also found partly intergranular fractures in mild steel. The importance of this is that because such material has a grain-boundary strength less than the cleavage strength, the temperature of transition from tough to brittle fracture is governed by the grain-boundary strength, and that if the grain-boundary weakness were removed an immediate substantial drop in transition temperature could be expected. However, in reply to a further question by Mr. Bardgett, we would certainly not wish to imply that " in commercial steels the effect of increasing grain size is merely to change the brittle fracture from cleavage to intergranular." The effect of grain size in commercial steels is probably complex, and it appears likely that the effects on yield stress and cleavage strength, about which we have information from the work of Petch,[12] are not the only considerations. The method of producing a particular grain size should also be considered as indicated by Mr. Bardgett in his question about the fine grain size produced in the 1% silicon alloy by cold working and annealing at 700° C. Longer times of heating at 700° C. have been tried and there is a progressive drop in ductility in tensile tests at −196° C. which indicates a rise in transition temperature in tension. Unfortunately

the results are complicated by the fact that at this temperature grain-coarsening occurs with time. This work is being repeated by investigating the effect of time of heating at a somewhat lower temperature after the initial treatment at 700° C., to try to avoid a change in grain size so that the diffusion effect might be isolated.

Careful reading of the text at the top of p. 95 should convince Mr. Bardgett that there is no contradiction with Table II. The fact that the tensile strength rises with decrease in temperature for the 1% Si alloy, is constant for the 2% Si alloy, and is lowered for the higher-silicon alloys is a consequence of the larger effect of silicon in raising the strength in the ductile condition compared with the brittle condition. We have stated that we think that the difficulty of cold-rolling the Fe–Si alloys is determined partly by the incidence of intergranular brittleness; we do not know whether there is any connection between their ability to be cold-rolled as thin sheet and any preferred orientation resulting from rolling.

Mr. Bardgett seems to have misunderstood the explanation given for the properties of the Fe–Ni alloys. The effect of tempering is explained in terms of two factors, one of which is the relief of stress in the α_2 structure and the other diffusion of impurities that would tend to decrease the grain-boundary strength. We do not associate the tendency to intergranular failure with the α_2 structure. It is presumed that the statement that nickel in commercial steels invariably lowers the transition temperature refers to quenched and tempered material in which the temper-brittle effect is avoided in the tempering treatment. This is not inconsistent with the present results in which the raising of the transition temperature by nickel is a grain-boundary effect; if this could have been suppressed, the effect of nickel in raising the cleavage strength might well have resulted in a lowering of the impact transition temperature in the tempered condition. However, there are several unknown factors which need investigating before the results for Fe–Ni alloys can be completely correlated with the behaviour of nickel in steels.

The question of 'star' fractures is interesting. It is true that they occur in Fe–Mn and Fe–Ni alloys having an α_2 structure. They have, of course, been reported in other materials, for example in steels transformed in the intermediate range.

The figures given by Mr. Bardgett for atomic diameter are misleading because they represent a variety of co-ordination numbers. We do not know how he arrives at his figure for manganese, which has a complex structure and is thus a difficult element to which to assign an atomic diameter; the figure of 2·75 does not appear in the reference he quotes. Leaving out manganese and taking atomic diameters for a co-ordination number of 8, the order for the difference between the atomic diameter of an element and that of iron is Mo, Cr, Si, Ni, which is different from the order given by Mr. Bardgett.

Professor Rudberg, in discussing the paper on Fe–N alloys, is correct in stating that the time of heating at 1000° C. in the experiments reported in Fig. 3 was longer than at 900° C. because of the slower rate of diffusion of nitrogen in γ iron at the higher temperature compared with that in α iron at the lower temperature, as can be seen from the footnote on p. 114. The results are apparently not affected by the transformation of austenite to ferrite on quenching, presumably because the austenite grain boundaries are also ferrite grain boundaries. The suggestion that the location of the ferrite boundaries makes itself felt in advance seems to be an inverted viewpoint. The position is surely that as ferrite forms from austenite, the austenite boundaries are natural sites for ferrite boundaries, although they are not the only ones if the ferrite grains are smaller than the austenite grains from which they grew. In any case, at 1000° C. we may well be getting towards the stage of random distribution of nitrogen atoms.

Professor Rudberg has made an interesting comment on the effect of grain size on the fracture stress. The fracture stress at −196° C. of one of the Fe–N alloys was less at 29·6 tons/sq. in. after furnace-cooling from 1150° C. to give a grain size of 3 grains/mm. than the 39 tons/sq. in. obtained after furnace-cooling from 950° C. which gave a grain size of 8 grains/mm. He considers an explanation of this in terms of grain-boundary concentration of nitrogen to be unlikely and asks whether the effect could be explained on grounds other than chemical ones. We agree that part of the explanation may be physical, similar to the effect of grain size on cleavage strength. However, there must be a concentration effect because there was only a trace of grain-boundary failure in the finer-grained condition and a distinct amount in the coarser-grained condition. This point merits further consideration.

From the contribution by Mr. Josefsson and Mr. Lagerberg it is difficult to avoid the conclusion that iron free from interstitial atoms should have a very low intergranular strength under all conditions. It remains to be proved that this is so. If the ideas put forward by Mr. Josefsson and Mr. Lagerberg were accepted, it would be difficult to explain the increase in brittle-fracture strength after quenching from temperatures above 700° C. The explanation in terms of the formation of nitrogen-rich austenite films at the ferrite grain boundaries is not convincing. It should be pointed out that the austenite films found by Tsou, Nutting, and Menter were formed on heating an ingot iron having carbide at the grain boundaries to a temperature within the range 600–700° C. where austenite of high carbon (and possibly nitrogen) content could form under equilibrium conditions. These films would not be expected to be retained at room temperature for higher temperatures of heating where the carbon and nitrogen contents of the austenite would be progressively reduced to the values represented by the composition of the iron. In any case it would be necessary to explain why the insertion of a film of austenite between two grains of ferrite should raise the strength of the grain boundary.

Nevertheless, we wish to emphasize that we regard the explanation of the grain-boundary weakness put forward in the paper as tentative and as being merely what we regard as the simplest available at present. We should like to record that in presenting the paper a modification was suggested which might make more acceptable the explanation of the recovery of brittle-fracture strength for temperatures of heating below about 500° C. Whereas in the paper we suggested a diffusion of nitrogen away from the grain boundaries at these temperatures, this is perhaps unnecessary, and a movement of nitrogen atoms along the boundaries to form clusters with iron atoms as a pre-precipitation stage could be sufficient effectively to denude the boundaries of nitrogen atoms and so break up a continuous nitrogen-rich region.

The objection of Mr. Josefsson and Mr. Lagerberg to the use of the parameter brittle-fracture strength is difficult to understand. The fact is that the load required to break in a brittle manner test pieces embrittled at 600° C. decreases as the temperature of test is lowered, as shown in Fig. 4. That this fact is hard to understand is no criticism of the use of the load (or the brittle-fracture stress derived from it) as a parameter in the assessment of results.

It is true that it is sometimes not possible to explain

the trends of transition temperature in the notch impact test in terms of yield stress at about this temperature and brittle-fracture strength, mainly because the brittle-fracture strength is only known at a temperature very much lower than the transition temperature in impact. The difficulty is not that the principle of the method is at fault but that the data available are insufficient.

An immediate reason why the idea of the development of a pressure of nitrogen gas in voids in the metal as suggested by Mr. Oliver is not likely to be helpful is that other systems (*e.g.*, Fe–P) show similar effects to the Fe–N alloys.

Mr. Bardgett's remarks on the Fe–N paper have been noted. We can do little other than speculate about the differences between the effects of carbon and nitrogen in iron.

Mr. **Josefsson** wrote: As in normal carbon steels, the transverse ductility is somewhat inferior to the longitudinal. This is reflected in the transition curves for transverse specimens (Fig. 9) and in the transition temperatures in Table VIII.

Ordinary tensile tests, in the longitudinal and transverse directions, carried out on 15-mm. plate from heat PF–1 (*see* Table VII) gave the following figures:

	Yield Strength σ_y, kg./sq. mm.	Ultimate Strength σ_u, kg./sq. mm.	Elongation , %	Reduction in Area ψ, %
Longitudinal	29·4	40·2	29·9	74·6
	29·3	39·8	28·2	75·7
Transverse	32·6	42·2	21·4	68·7
	31·8	41·7	23·7	69·4

The carbon limit (0·015%) is based purely on impact transition behaviour, the transition temperatures starting to rise when the carbon percentage exceeds this figure. The occurrence of cementite is not exclusively governed by the carbon content, but also depends upon cooling rate and ferritic grain size. After slow cooling, some grain-boundary carbide is also present in steel PF–1 with 0·005% C. Even after normalizing many of the pearlite-free steels investigated contain small amounts of grain-boundary carbide.

It is true that nitrogen is converted to an insoluble nitride in titanium-containing steels. But aluminium has the same effect, and therefore steels *A* and *B* are both practically free of nitrogen in solid solution. The difference is that steel *A* (but not steel *B*) contains carbon in solid solution.

The welding electrode KB 5225 is of the low-hydrogen basic-coated type, corresponding to the English *E*614.

In his first point, M. Crussard suggests that there is a different mechanism for brittle fracture in the decarburized specimens from that in those containing grain-boundary carbide. We did not investigate this question in detail, because neither the macroscopic appearance of the fractured specimens nor the microscopical examination of sections of the fracture surfaces hinted at any differences in the fracture history for these two materials. I agree, however, that if greater amounts of grain-boundary carbides are present, this may affect the fracture mechanism; in this case, however, the character of the transition curve is generally changed from the very steep type in Figs. 2 and 3 into a more gradually sloping one.

From the interesting points about fracture initiation, raised by M. Crussard, it is possible to obtain an alternative (or additional) cause for the good ductility of the pearlite-free steels, in spite of their high content of phosphorus and nitrogen. If the crack initiation by local ductile fracture is a major cause for brittleness, local variations in deformation resistance will be critical for the ductility, as the very soft parts of the structure can be plastically deformed (until the start of a local shear fracture) at rather low stress levels, where the main part of the structure still behaves elastically, which may initiate a low-energy, apparently brittle fracture. Now, as pointed out in the paper and shown in Table III, the pearlite-free steels are characterized by a remarkable uniformity in composition; this seems to be true also on a microscopic scale, judging from the uniform microstructure of equiaxed ferrite grains. From this one might expect a relatively high uniformity with respect to deformation resistance, and therefore less tendency to fracture-inducing local deformation.

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Correspondence on the Paper—

DEEP-DRAWING PROPERTIES OF SHEET STEEL. FUNDAMENTAL PRINCIPLES AND TEST METHODS*

By Olov Svahn

Mr. **G. Murray** (*Chairman*) and Dr. **B. B. Hundy** wrote, on behalf of the Cupping-Press Sub-Committee of the Mechanical Working Division of B.I.S.R.A.: Workers in at least three countries (Sweden, Germany, and the U.K.) are actively engaged in attempting to find a test to determine the suitability of metal sheet for deep-drawing. So far, it seems that the findings of workers in these countries are in general agreement: tests (iii), (iv), and (v) on p. 129 of Dr. Svahn's paper are not usually suitable for classifying sheet according to its suitability for deep-drawing, and it is only for drawing operations where stretching is most important

* *J. Iron Steel Inst.*, 1954, vol. 177, Mar., pp. 129–142.

that the Erichsen test is a reliable guide. The usefulness of some form of cupping test (using the maximum blank size that will draw satisfactorily as a criterion) seems to be recognized by all, and it seems eminently desirable that closer contact should be established between the various groups of workers in these different countries. It might then be possible eventually to recommend a standard form of cupping test, and the possibility of international joint development work leading to this should be seriously considered.

The Sheet Metal Working Committee of the British Iron and Steel Research Association have been considering the problem of a deep-drawing test for some years, and have devoted most of their attention to the test proposed by Professor Swift before the war.[1] A sub-committee to consider the Swift cupping test in more detail was set up by the parent Committee in 1951. In its present form, the press used for this test produces a 2-in. dia. cup, and the blank holding pressure can be varied from 5–10 to about 700–800 lb./sq. in.; early versions of the press were designed for constant-clearance blank holding, but this method was found to be not always satisfactory, owing to the difficulty of obtaining sufficient rigidity in the press. The press has two punches, one with a hemispherical head and the other with a flat head ($\frac{1}{4}$-in. edge radius). Tests[2] by Professor Swift and his colleagues have shown that sheet materials can be classified in different ways by these two tools; it is thought that the results obtained using the hemispherical punch may be useful for classifying material for drawing operations involving a moderate amount of stretching, whilst the flat-headed punch may be more suitable for operations approaching pure deep-drawing. The dies used give a clearance of about 30% of the sheet thickness and have a $\frac{1}{4}$-in. edge radius.

Four nominally identical models of this press were built by the same manufacturer and, of these, one is used in the B.I.S.R.A. laboratories at Sheffield and the others are in works' laboratories. Most of the work to date has been devoted to obtaining mutual consistency between the four presses. It has been found that minor differences in the profile of the tools (especially any bad blending-in of the radii) and in the surface finish can lead to a marked lack of consistency. The lubricant used is also very important, and it has been found that altering the lubrication profoundly affects the maximum blank size that can be drawn. The Sub-Committee now consider that, provided that a standard testing procedure is adopted and the tools are in good condition, consistent results can be obtained between the four presses, and the results are reproducible. The Sub-Committee are satisfied that the Swift test carried out on a press as outlined here is reliable, but they do not consider that this or any other cupping test is, at present, suitable as a standard for testing the drawing qualities of sheet metal. They are rather surprised, in view of the data given in Table II of Dr. Svahn's paper, that the Jernkontoret Committee feel that they have reached a stage in their work where they can recommend a cupping test as an international standard. The B.I.S.R.A. Cupping-Press Sub-Committee feel that, in the light of their own work and that of Dr. Svahn, further experience with the test in the press shop itself is required.

They consider that tests should be carried out using these presses (with both the hemispherical and the flat-topped punches) on a wide variety of sheet metals used in the press shops, and that the results of these tests should be correlated with the behaviour of the sheets when used for a variety of different pressings in the press shops. It is felt that more experience in the use of the press and in the interpretation of the results must be gained before it is possible to decide whether it can be used satisfactorily as a deep-drawing sheet acceptance test.

The most disturbing feature of Dr. Svahn's paper is Table II, which shows that there is little correlation between various cupping tests. Some light may be thrown on this if Dr. Svahn could comment on the following questions:

(i) Were the steel sheets temper-rolled or not? If some of the sheets were temper-rolled, strain ageing between one set of tests and another might lead to observed differences in behaviour.

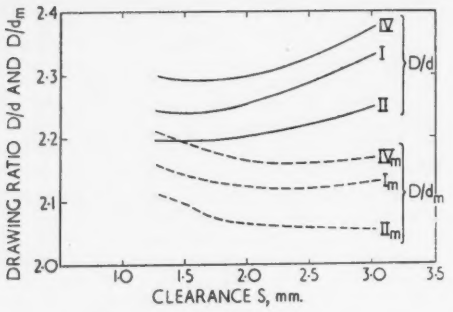
(ii) How were the blanks for the cupping tests prepared? We have found that the blanks should be machined to size on a lathe and any flash must be removed by smoothing with emery paper if consistent results are to be obtained. Punching out the blanks is not satisfactory unless the sheared edge of the blank is afterwards removed by machining on a lathe.

(iii) What were the conditions of lubrication of the blank? We have found that it is essential to standardize exactly the lubricant and the method of application, and it appears that Dr. Svahn has obtained similar results.

(iv) Did the condition of the tools vary at all during the tests? Our experiments have shown that pick-up on the die can affect the results and that touching-up of the die to remove this has to be done very carefully so as to maintain a good surface finish and good blending of the radii. Dr. Svahn seems to be aware of this, and refers to the importance of surface finish.

Dr. Svahn gives some figures for the correlation of the KTH tests (June, 1946), with the usual tensile properties; it would be interesting to learn the degree of correlation between the ranking as determined by these properties and that determined by the other cupping tests (Table II suggests that it would not be very good). A criterion of drawing quality which is sometimes used but which Dr. Svahn has not considered is the ratio of the yield stress to the ultimate tensile strength. On this basis the ranking order of these steels seems to be ADCBIJH. In many press shops this criterion is generally considered in conjunction with the uniform and total elongations, and on this basis the ranking order might be expected to be ACBDHIJ, which gives rather more reasonable correlation, at least with the KTH tests (June, 1946).

It is interesting to note in the second part of the paper that Dr. Svahn's findings on the various factors affecting the possible depth of draw are in general agreement with the earlier work of Professor Swift (two reviews of Professor Swift's work on deep-drawing were published recently[3,4]). This fundamental work is most useful in interpreting some of these effects of deep-drawing variables, and it is a pity that Dr. Svahn has not cited any of Professor Swift's work after 1947. His work definitely shows that there can be no single cupping test for determining the deep-drawing quality of metal sheet for producing different shapes, as the depth of draw (maximum blank diameter that can be drawn) depends markedly on the tool geometry. Thus, by suitable choice of the punch profile, it is possible to draw a larger blank of half-hard aluminium sheet than of soft aluminium, whilst with other punch profiles it is possible to draw a larger blank in soft aluminium than in half-hard. The function of the lubricant in drawing flat- and round-bottomed cups is very important and has been discussed recently by two of Professor Swift's colleagues,[5] at the Symposium on Metal Working Oils held by the Institute of Petroleum. As mentioned earlier, the B.I.S.R.A. Sub-Committee



D = Limiting blank dia.
d = Punch dia.
d_m = Mean cup dia. ($= d + S$)

Fig. A—Drawing ratio as a function of clearance between punch and die

hope to overcome this difficulty of ranking sheet metal for a deep-drawing operation by the use of the two differently shaped punches.

Mr. **D. A. Barlow** (Aluminium Laboratories Ltd.) wrote: In Fig. 11 of Dr. Svahn's paper, curves are given for three steels showing at first a decrease and then an increase in drawability with increasing clearance between punch and die. Drawability is here measured as the maximum ratio of (blank dia.)/(punch dia.), which will not cause rupture of more than 10% of the cups, but a more accurate measure of the deformation received by the cup is (blank dia.)/(mean cup dia.). In this case, Dr. Svahn has used a constant punch diameter and has obtained the various clearances by using dies of various diameters. With the larger clearances in particular, the cup assumes a tapered shape and the mean cup diameter is approximately $\frac{1}{2}$ (die dia. + punch dia.) or (punch dia. + radial clearance). Dr. Svahn's curves are replotted in Fig. *A* on this basis, and it will be seen that the real drawability is almost constant with radial punch-die clearances of over 2 mm. At smaller clearances, the drawability increases, perhaps owing to the higher strength of the cup walls when fitting more closely to the punch. At still smaller clearances, drawability will decrease, owing to excessive ironing loads.

AUTHOR'S REPLY

Dr. **Olov Svahn** (Sandviken Jernverks AB) wrote: The steel sheets were not temper rolled. During the research work the ageing tendency of the materials was under very intensive observation. A certain strain ageing was found but cannot explain the lack of correlation between the test series. The only way to explain this is to examine the test methods themselves and the influence of different factors on the tools and their function.

The blanks are first punched; then the edge of the blank is removed by machining, turning, or grinding. At least 0·5 mm. has to be taken away on the radius and the diameter has to be machined with a tolerance of $\pm$ 0·05 mm.

The lubrication is as mentioned very important. We have standardized on a high sulphur, high chlorine oil, very suitable for deep drawing. The condition of the oil has been examined from time to time and we have found that an ageing process during the summer will influence the results. These circumstances have been taken into consideration. We have also discussed whether another oil would influence the grouping of, for instance, brass and aluminium in the correlation series. The results are reproduced very well with different types of oil, even if the nominal values of the drawability are not the same. The application of the lubrication has also been standardized in all cases.

The tools naturally varied during the tests; this was unavoidable. The amount of variation during the tests was studied by working with one of the materials as a base standard; tests have been repeated with one of the materials tested when the tools were newly polished. The tools are inspected after every test drawing and as soon as the pickup and other defects in the working surfaces are noticeable a calibrating test has been made and the tools repolished. The surface finish has been tested every time.

The proposal that the usual tensile properties should give a better degree of correlation compared with the deep-drawing test is right in so far that the true-stress/true-strain diagram shows a very good correlation factor. However, not only the correlation factor has to be examined, but also the significance of a correlation found in one test series. As far as the KTH tests (June, 1946) are concerned, the significance is not acceptable. The explanation of this needs discussing.

I agree that it is impossible to lay down any general cupping test for determining the deep-drawing quality of metals for all different shapes of drawing details. However, the Swedish committee has divided the tests into two general groups, as given in my paper. This differentiation is the first step towards a better correspondence with practice. It is not certain that one of these testing methods can give perfect correlation for a complicated deep-drawing operation in a certain material and of a special form. If the intention is to find a testing method or test acceptable to producers and consumers with the following aims

(i) To ascertain a given figure of drawability (*e.g.*, as a delivery test)
(ii) To find the maximum drawability of a particular sheet compared with other sheets,

it would be better to have a definite test that is generally acceptable than to use the existing tests, which are inaccurate and unreproducible.

In answer to Mr. Barlow, it is true that the relation (blank diameter)/(mean cup diameter) will be a more accurate figure for the grade of deformation in the material than the ratio (blank diameter)/(punch diameter). However, in a practical test it is necessary to calculate with some absolute values on the tool used for the test. Then it is immaterial whether the punch diameter or the die diameter is used. The tests shown in Fig. 11 are designed to show whether the grouping of the materials is the same for different clearances. The method of calculating the degree of deformation has only an academic interest.

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2. H. W. Swift: Private communications.
3. J. Willis: " Deep Drawing. A Review of the Practical Aspects of Professor H. W. Swift's Researches ": 1954, London, Butterworths Scientific Publications.
4. H. W. Swift: *Sheet Metal Ind.*, 1954, vol. 31, p. 817.
5. E. M. Loxley and P. Freeman: *J. Inst. Petroleum*, 1954, vol. 40, pp. 299–307.

IRON AND STEELWORKS ENGINEERING

THE IRON AND STEEL ENGINEERS GROUP

THE IRON AND STEEL ENGINEERS GROUP

REPORT OF THE TWENTY-SIXTH MEETING

THE TWENTY-SIXTH MEETING OF THE IRON AND STEEL ENGINEERS GROUP of The Iron and Steel Institute was held, by invitation of Mr. H. BOOT, Managing Director, Consett Iron Co., Ltd., in Newcastle-upon-Tyne from Monday, 13th September, to Thursday, 16th September, 1954. Mr. W. M. LARKE (Stewarts and Lloyds, Ltd.), Chairman of the Group, occupied the Chair.

The Technical Sessions at the Royal Station Hotel were devoted to the presentation and discussion of seven papers by members of the staff of the Consett Iron Co., Ltd., on the developments at Consett, followed by visits to the Works at Consett and Jarrow and to the Ore Discharge Terminal at Tyne Improvement Commissioners Quay (by kind permission of the Tyne Improvement Commission).

At the MORNING SESSION on Tuesday, 14th September, the following papers were presented and discussed: " The Duplex Steelmaking Process," by Mr. W. E. WARD (Works Production Manager); " The Slabbing/Blooming and Continuous Billet Mills: Installation and Layout," by Mr. A. MARSHALL (Development Engineer), and " Operational Experiences with the Slabbing/Blooming and Continuous Billet Mills," by Mr. A. HENDERSON (Mill Manager), the second and third papers being discussed jointly.

At the AFTERNOON SESSION the first three papers were discussed jointly: " Electrical Developments at Consett," by Mr. T. COXON (Consulting Electrical Engineer), " Some Aspects of Electrical Maintenance," by Mr. R. MATHIESON (Chief Maintenance Engineer (Electrical)), and " The Operation of an Industrial Power Station at Consett Iron Works," by Mr. T. F. HUTCHINSON (Power Station Manager). The final paper discussed was " Imported Ore—Tyne Dock to Consett. Description of and Experiences with New Ore Handling Plant," by Mr. S. C. PEARSON (Transport Manager).

Abstracts of the papers, together with a short report of the discussion on them, are printed below. A fuller report will be published in the Institute's Special Report Series. Copies of the papers were distributed to Members attending the Meeting, and are available on loan from the Institute Library. A Technical Survey of the Consett Works appeared in a Special Issue of *Iron and Coal Trades Review* in March, 1954.

ABSTRACTS OF PAPERS

The Duplex Steelmaking Process

By W. E. Ward

The acid Bessemer converter plant, commissioned in January, 1954, forms an integral part of the open-hearth plant and is housed in the existing melting shop. It consists of two 25-ton converters with all-welded shells. Hot metal with the following average composition: Si, 1·2%; P, 0·6–1·0%; S, 0·04–0·07%; Mn, 0·5% is used. The blown metal is transferred to the open-hearth furnaces by ladle, and charging begins immediately after fettling. A heat of 125 tons of blown metal, 35 tons of hot metal, and 25 tons of scrap was found to be the optimum for increase in output and working rate, although the greatest possible speed in working the duplex charge is essential for economic operation.

The Slabbing/Blooming and Continuous Billet Mills: Installation and Layout

By A. Marshall

The mill comprises a stripper bay building, a soaking pit building, and the main mill building, consisting of two bays over 1000 ft. long. Three batteries of 4-cell soaking pits equipped with Escher-type air and gas recuperators are installed.

The slabbing/blooming mill is a 44 in. × 108 in. reversing mill, driven by twin 4750-h.p. motors, with auxiliaries including ingot transfer car, roller tables, ingot weighing machine, transfer banks, and a slab shear. It is capable of slabbing 20-ton ingots and blooming 10-ton ingots.

The continuous billet mill comprises a 6-stand roughing mill, capable of producing 9–3¼-in. square billets and slabs 24–9 in. wide down to 2½ in. thick, and a 4-stand

finishing mill, producing billets from 3½ to 1¾ in. square and small slabs from 8 in. × 2 in. down to 4 in. × 1½ in. Auxiliary equipment includes a 220-ft. broadside transfer bank, a 66-in. dia. hot saw, and an electric flying shear.

Operational Experiences with the Slabbing/Blooming and Continuous Billet Mills

By A. Henderson

The slabbing/blooming mill produces slabs for re-rolling in the plate mills, slabs and blooms for sale, and blooms for the continuous billet mill. The products of the latter are used for serving the strip and section mills, whilst a large tonnage is sold.

The soaking pits require very little maintenance; brickwork and maintenance costs each average about 1*d.* per ton. Blast-furnace gas is used for heating, 12 therms/ton being obtained.

The estimated life of the cast and forged steel rolls of the slabbing/blooming mill is about 350,000 tons. In the billet mill alloy rolls and alloy rolls with a steel base are used, and roll lives are satisfactory.

Electrical Developments at Consett

By T. Coxon

The power station is also a turbo-blower station serving the blast-furnaces. The installed generator capacity of 45 MW. is adequate for present and immediate future requirements; the switchgear and transformers selected give the highest degree of reliability and flexibility in the control and distribution systems.

The electricity supply to the blast-furnaces, coke ovens, steelworks, rolling mills, and Bessemer plant incorporates many outstanding design features, all intended to increase efficiency, lower costs, and simplify maintenance and operation.

(Fuller details are given in a paper by Mr. Coxon, published in *J. Iron Steel Inst.*, 1950, vol. 164, pp. 327–348).

Some Aspects of Electrical Maintenance

By R. Mathieson

Planned maintenance depends on the proper use and application of plant records, inspection schedules, and history cards. It can only be successful, however, if the personnel responsible for carrying it out are carefully selected and supervised. When new plant is installed, the maintenance staff must be fully conversant with its operation, and a series of lectures was held at Consett to provide this information about the new plant.

A biennial survey of all electrical equipment at Consett is carried out by an impartial outside source, and has proved to be of the greatest value.

The Operation of an Industrial Power Station at Consett Iron Works

By T. F. Hutchinson

The power station consists of five boilers raising steam at 405 lb./sq. in. and 400° C., three of which can burn blast-furnace gas, coke-oven gas, or pulverized coal, singly or in combination.

The power plant comprises three 15-MW. 11-kV. turbo-alternators, capable of passing out 80,000 lb. of steam per hr. at 160 lb./sq. in. There are two turbo-blowers rated at 45,000 cu. ft. of air per min. at 30 lb./sq. in. Three 750-kW. rotary converters take A.C. at 11 kV. and convert it to D.C. at 500 V. Electric power is distributed at 11 kV., 3·3 kV., and 500 V. (A.C.) and 500 V. (D.C.).

Imported Ore—Tyne Dock to Consett

By S. C. Pearson

Negotiations with the Tyne Improvement Commission and the British Transport Commission led to the building of a new wharf and a deep-water quay capable of taking ships up to 25,000 tons.

The ore-discharging plant there consists of five 10-ton electric travelling grabbing cranes, which can handle 300 tons/hr. Two distribution conveyors take the ore to ten 13,000-cu. ft. steel storage bunkers, from which it is loaded on railway wagons by two electric travelling weigh-hoppers.

The ore-handling plant at Consett, which is capable of discharging and handling over 10,000 tons/day, includes bunkers, travelling feeders, conveyors, and stocking and reclaiming bridges. The 300-ft. span ore bridge is the largest outside the U.S.A.

DISCUSSION

Duplex Steelmaking

In the discussion on Mr. Ward's paper, favourable comments were made about the advantages of using the duplex process, which increased the output from the open-hearth plant without any deterioration in the quality of the steel made (although the phosphorus content might still be reduced). The rather high blast volume at Consett of 35,000 cu. ft./min. was justified by the saving in time. The need for the application of operational research to the process was emphasized.

The Slabbing/Blooming and Continuous Billet Mills

The discussion was devoted largely to an assessment of ' teething troubles ' experienced on the new mill. The early inconsistency of the flying shear had been examined and was now almost fully rectified. The problem of disposing of scale at the cogging rolls was solved by the introduction of grids underneath.

Knurling was suggested as a means of extending the life of cogging rolls, but the mill had not been in operation long enough for an accurate figure for roll life to be given. The use of oil-film roll bearings and of welding were defended.

The Electrical System at Consett

Some doubt was raised about the economics of installing equipment producing 45 MW. as compared with purchasing the supply from the B.E.A. At Consett departments paid at the same rate as would be demanded by the Area Board, and, with a total cost of generation of ¾*d.* per unit, the economy was sound.

The relative merits of Ward-Leonard and D.C. resistance control for skip hoists were considered. The deciding factor was that of the size of the skip hoist.

The problem of magnetic dust accumulating between the magnet and armature of the Igranic brake could be remedied by more frequent maintenance, but the period between cleanings could be increased by fitting a simple seal; elastic adhesive tape made an efficient seal.

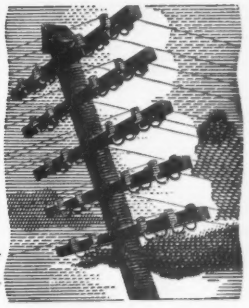
The Consett system of having a special team for crane maintenance was approved, but it was suggested that the main reason for crane breakdowns was shock; shock absorbers might reduce the maintenance needed on cranes.

Ore Handling

Two points arose in the discussion: the need for larger, better designed ore ships, and for a quicker turnround of ore trains, made up of larger wagons. Previously there had been a reluctance to build large ore ships, owing to the inadequate port facilities in the U.K. These larger ships should be designed to facilitate discharging; present unloading times could be drastically cut. To ensure an even flow, ship arrivals should be made more regular.

The railway haul from port to works was not yet as efficient as it might be. The 56-ton ore wagon had speeded up the transfer of ore, but their tare weight might be reduced substantially. Once again, the flow of materials needed to be more regular, and the provision of ample bunker storage was very important.

N E W S



ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE
President-Elect

At the Autumn General Meeting of the Institute, held on 17th and 18th November, 1954, it was announced that the Council had nominated Sir Charles Bruce-Gardner, Bt., M.I.Mech.E., for election as President at the Annual General Meeting, 1955, and to hold office for one year.

Meetings in 1955

The **Annual General Meeting** of The Iron and Steel Institute will be held on **Wednesday and Thursday, 27th and 28th April, 1955,** and the Dinner for Members will be held at Grosvenor House, Park Lane, London, W.1, on 27th April. The **Autumn General Meeting** will be held on **Wednesday and Thursday, 16th and 17th November.**

The Institute has accepted an invitation from the Lincolnshire Iron and Steel Institute to hold a **Special Meeting in Scunthorpe from 12th to 14th October.**

Joint Metallurgical Societies' Meeting in Europe, 1955

A joint meeting between the American Institute of Mining and Metallurgical Engineers, the American Society for Metals, and leading metallurgical societies in Great Britain, Germany, Belgium, and France will be held in Europe on 1st–18th June, 1955. Details of the meeting were given in the December, 1954, issue of the *Journal*, p. 396.

Andrew Carnegie Research Fund

The following awards have been made:

Mr. A. S. Appleton (Liverpool University): £150 to assist his research on the kinetics of graphitization in high carbon, Fe–C, and Fe–Co–C alloys, together with subsidiary diffusion experiments on Fe–C–Si and the above alloys.

Mr. C. J. Ball (Queens' College, Cambridge): £410 to assist his research on the plastic deformation of metals.

Mr. P. R. Vaughan Evans (University College, Swansea): £350 to assist his research on the mechanical properties and internal stresses of various strip steels after various isothermal treatments.

Mr. J. Wilcock (Liverpool University): A second grant of £50 to assist his research on the influence of manganese, sulphur, and the sulphur to manganese ratio on the kinetics of first-stage graphitization.

NEWS OF MEMBERS

▶ Dr. Maurice Cook, Joint Managing Director of the Metals Division of Imperial Chemical Industries Ltd., has been elected Chairman of the British Non-Ferrous Metals Research Association.

▶ Dr. O. W. Ellis has retired from his appointment as Director of the Department of Engineering and Metallurgy at the Ontario Research Foundation, Toronto.

▶ Mr. M. M. Hallett has been appointed Managing Director of Chamberlin and Hill Ltd., Walsall.

▶ Mr. J. Heslop has taken up an appointment with the Royal Naval Scientific Service at Rosyth.

▶ Mr. P. S. Mathur is now working in the blast-furnace department at the Burnpur Works of the Indian Iron and Steel Co., Ltd.

New Year's Honours

▶ Mr. E. W. Senior, Commercial Director of the British Iron and Steel Federation, has been created a C.M.G.

▶ Mr. L. Chapman, Managing Director of Wm. Jessop and Sons, Ltd., has been created a C.B.E.

▶ Mr. G. D. Elliot, Works Manager (Iron—Blast Furnaces), Appleby-Frodingham Steel Co., has been awarded the O.B.E.

▶ Mr. Paul List, M.B.E., of Luxembourg, has been awarded the O.B.E.

▶ Mr. D. R. Wattleworth, Director and General Works Manager, Workington Iron and Steel Co., has been awarded the O.B.E.

▶ The Secretary of the Institution of Mining and Metallurgy, Mr. W. J. Felton, has been awarded the O.B.E.

CONTRIBUTORS TO THE JOURNAL

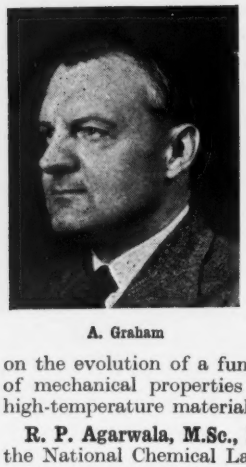
M. D. Ashton, B.Sc.—Scientific Officer in the Fuel Technology Section, Physics Department of the British Iron and Steel Research Association.

M. D. Ashton

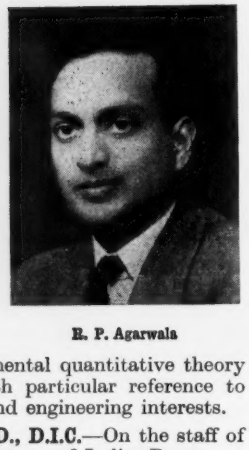
Mr. Ashton graduated in Chemistry from the Northern Polytechnic in 1948, and in the following year gained a post-graduate diploma in Chemical Engineering at University College, London. He spent two and a half years with Standard Telecommunication Laboratories Ltd., where he was engaged in research and development work on plastic materials, before taking up his present appointment in 1952.

A. Graham, M.Sc.—Principal Scientific Officer at the National Gas Turbine Establishment, Farnborough.

Mr. Graham was born in London in 1910, and was educated at the Central Foundation School, University College, and at the Imperial College of Science and Technology, London. He was awarded the degree of B.Sc. with 1st Class Honours in Physics, and later obtained the degree of M.Sc. for research on the instability of heated fluids. From 1932 to 1944 he was engaged in research on various experimental and theoretical problems for the Royal Aircraft Establishment, working in the Aerodynamics, Air Defence, and Armament Departments. He was then transferred to Headquarters at Farnborough, where he remained until he took up his present appointment in 1947. He has since been working



A. Graham



R. P. Agarwala

on the evolution of a fundamental quantitative theory of mechanical properties with particular reference to high-temperature materials and engineering interests.

R. P. Agarwala, M.Sc., Ph.D., D.I.C.—On the staff of the National Chemical Laboratory of India, Poona.

Dr. Agarwala obtained the degrees of B.Sc. (1948) and M.Sc. (1950) of Agra University, India. He came to the U.K. in 1951 and became a Wakefield Research Scholar at the Imperial College of Science and Technology in the following year. In 1953 he was awarded a D.I.C. and the degree of Ph.D. of London University. He then returned to India to take up his present research post.

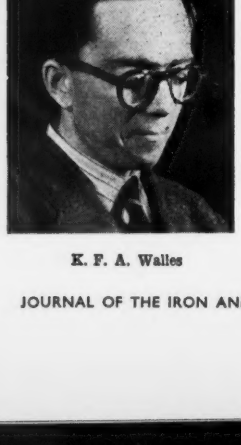
K. F. A. Walles, B.A.—Experimental Officer at the National Gas Turbine Establishment, Farnborough.

Mr. Walles was born at Southampton in 1925. He was awarded a State Scholarship to Cambridge University and a Christ's College Exhibition, and obtained the degree of B.A. in Natural Sciences in 1945. For the next year he worked in the Armament and Metallurgy Departments of the Royal Aircraft Establishment, and then took up a post in the Physics Laboratory of Metropolitan-Vickers Electrical Co., Ltd. He took up his present appointment in 1948, and is now engaged on research into the mechanical properties of materials at high temperatures.

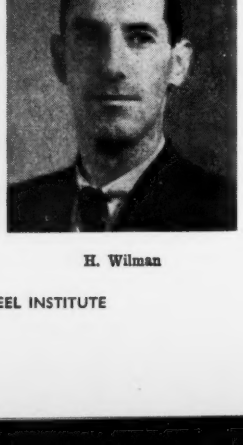
H. Wilman, D.Sc., Ph.D., D.I.C., F.Inst.P.—In charge of the Applied Physical Chemistry Laboratory, Imperial College of Science and Technology.

Dr. Wilman studied at the Royal College of Science, and in 1932 was awarded the A.R.C.S. and the degree of B.Sc. (Hons.) in Physics. For the next two years he carried out research using the electron diffraction technique under Professor G. I. Finch in the Chemical Technology Department of Imperial College, and was awarded the D.I.C. and the degree of Ph.D. of London University. He continued this research and in 1937–38 acted as assistant to Professor Finch at Brussels University. From 1938 to 1940 he held the Moseley Research Studentship of the Royal Society. During the war he carried out research on the production of Cs–Cs_2O–Ag infra-red photocathodes for the Admiralty and made a detailed study of their structures and optical properties.

In 1946 Dr. Wilman became Lecturer in Applied Physical Chemistry at Imperial College, and continued electron diffraction research and the study of surface structure and properties. He was awarded the degree of D.Sc. in 1951. Since Professor Finch left the College in 1952 he has been in sole charge of the work on surface structure. He is the author of many papers on subjects including electron diffraction techniques, determination of molecular structure, crystal growth and orientation, structure and growth of condensed and electrodeposited layers, oxidation of metal surfaces, intercrystalline boundaries, and deformation of crystals.



K. F. A. Walles



H. Wilman

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Twenty-Second Blast-Furnace Conference

The 22nd Blast-Furnace Conference was held at Ashorne Hill, Leamington Spa, on 9th and 10th November, 1954. The Chairmanship of the Conference was shared jointly by Mr. R. A. Hacking and Mr. G. H. Johnson.

During the first session, which was devoted to blast-furnace refractories, Mr. H. M. Richardson introduced a survey of the results of the examination of the brickwork of blown-out blast-furnaces (Special Report No. 51 in the series published by The Iron and Steel Institute). From the examination of 79 furnaces at 30 plants, the various findings had been assembled and formed a basis for analysing the various types of brickwork deterioration from the top of the stack to the hearth. Dr. Rigby, in presenting a paper on present trends in blast-furnace linings, continued the story by describing present-day developments in blast-furnace design and construction and attempted to forecast future work. He referred to the increase of furnace size, the use of normal, super-duty, and carbon refractories, water-cooling of the bosh and hearth, and weakening of the stack by carbon monoxide disintegration. To obtain a longer furnace life, attempts would be made to develop materials to an agreed specification for furnace use.

The discussion was opened by Dr. H. R. Lahr, who gave details of high-fired super-duty bricks and described their excellent after-contraction properties. Reference was also made to the construction of the lining, and it was emphasized that it was desirable to have as few joints as possible and that, as the cost of a lining was small per ton of pig iron produced, there was ample margin for obtaining a brick of improved performance. The possibility of other materials being used for linings should not be overlooked.

Mr. J. W. Houghton opened the second session with a paper on the burdening of the blast-furnace for slag composition. He referred to the work of Osborne and others on properties of mixtures in the CaO–MgO–Al_2O_3–SiO_2 system, and indicated the optimum composition for blast-furnace slags with maximum desulphurizing potential and minimum viscosity for practices used in this country. During discussion, members gave details of the slag compositions required for their iron specification. At one plant, to obtain the necessary desulphurization, it had been necessary to use high-magnesia slag, thus confirming the American theories.

At the third session a complementary paper given by Mr. W. Banks reported experiments on high-alumina, low-silica operation. These had taken the form of trials for a short period when smelting Conakry ore. It had

been shown, in support of the results obtained by Osborne, that slags of both 15% and 35% alumina could be used, provided that the other components were suitably adjusted. In discussion it was pointed out that Conakry ore could become an important source of iron and, if it were desirable, burdens containing Conakry ore as the major constituent could be used by running these high-alumina slags. Reference was also made to the commercial possibilities of disposing of high-alumina slags; the specification for high-alumina cement could be met, and this might prove a useful outlet.

At the final session, the effect of the speed of relining and blowing-in on the life of blast-furnace linings was considered. Mr. G. D. Elliot presented a paper outlining the important aspects of rapid relining and the conditions occurring in the hearth during blowing-in. It was stressed that a great deal more information was required before the effect of the various rates of blowing-in could be critically assessed. In a second paper Mr. Corke gave constructional and blowing-in details where drying-out and drafting of the furnace had occurred. A full record of the operating conditions for the blow-in were given for a period of three weeks of a furnace campaign. Mr. Lakin gave an account of laboratory experiments on a loaded refractory wall 3 ft. × 2 ft. × 22½ in. which was subjected to different conditions of flame impingement on one face.

The discussion was opened by Mr. Richardson who gave details of the performance of bricks tested in the laboratory under conditions similar to those occurring in the blast-furnace. Various methods of relining and blowing-in were described by operators; the manufacturers of refractories also contributed their opinions and drew comparisons between the initial lining conditions of kilns and other furnaces and conditions in the blast-furnace.

AFFILIATED LOCAL SOCIETIES

Lists of Officers for the session 1954–55 are as follows:

Newport and District Metallurgical Society

President: G. H. Latham, LL.D., J.P.

Hon. Secretary: T. G. Grey-Davies

Hon. Treasurer: J. E. Wells

Committee: C. I. F. Mackay (*Chairman*), H. R. Davies, E. D. Harry, B.Sc., B. Jones, D.Sc., C. J. Morgan, R. A. Owen-Barnett, M.Sc., H. T. Roberts, H. H. Stanley, J. G. Vinsen, B.Sc.

Liverpool Metallurgical Society

President: G. T. Callis, B.Sc.

Vice-Presidents: A. E. Griffin, T. H. Fletcher

Hon. Secretary: S. V. Radcliffe, B.Eng.

Hon. Treasurer: R. S. Moore

Committee: W. I. Govier, S. J. Kennett, B.Sc., Ph.D., E. C. Rollason, M.Sc., Ph.D., W. Rose, B. Todd, B.Eng., A. Ward

Committee (ex officio): R. S. Brown, M.B.E., H. Edwards, B. L. Farthing, R. J. Miller, M.Sc.

Ebbw Vale Metallurgical Society

President: J. J. Beese

Vice-President: J. F. Smith

Hon. Treasurer: T. Jones

Joint Hon. Secretaries: F. W. Lewis, W. B. Williams

Committee: W. H. R. Bird, M.A., B.Sc., W. J. Davies, J. K. Edwards, M.A., P. Edwards, W. Goatman, I. Giddings, W. D. Jenkins, B.Sc., F. Llewellyn, A. W. Ellis, J. T. Mills, H. F. Padbury, G. R. Parkes, Ph.D., F. E. Probyn, R. C. Powell, B.Sc., J. Thickins, B.Sc., W. E. Williams.

Swansea and District Metallurgical Society

President: Prof. H. O'Neill, D.Sc., M.Met.

Vice-Presidents: R. W. Evans, M.Met., R. Walker

Hon. Treasurer: U. Williams

Joint Hon. Secretaries: H. A. Cooke, M. J. A. Thomas, B.Sc.

Council: J. S. Caswell, M.Sc. (*Chairman*), R. G. Davies, M. L. Hughes, Ph.D., A. J. K. Honeyman, B.Sc., D. A. Bishop, R. J. Thomas, E. F. Meyer, W. R. Hitchings, R. Higgins, O.B.E., D. G. Jones, B.Sc.

Cleveland Institution of Engineers

President: H. M. Morgan

Vice-Presidents: A. Dorman, E. W. Jackson

Hon. General Secretary and Treasurer: F. G. E. Webster

Hon. Technical Secretary: L. Stephenson

Members of Council: N. Blaney, H. H. Utley, T. W. Thursfield, W. R. Brown, E. Drewery, T. W. Hurst, E. F. Brown, E. Easson, T. H. Stayman, J. Johnston, F. B. George

Members of Council (Past Presidents): A. Scholes, J. H. Patchett, C. Moody, E. T. Judge, G. B. Thomas, N. C. Lake, F. Kennedy.

Staffordshire Iron and Steel Institute

President: K. G. Lewis, M.Sc., M.Sc.(Tech.)

Vice-Presidents: A. Biddulph, G. Parkin

Hon. Secretary: H. J. E. Jones

Hon. Treasurer: G. E. Lunt

Members of Council: W. Angus, J. A. Davies, G. T. Hampton, L. W. Law, R. B. Lane, M. R. Liddiatt, V. L. Nicholls, B. V. Peters, N. F. Truman, W. M. S. Walker, M.Sc., Ph.D., W. H. Wesson, F. V. Wright

Members of Council (Past Presidents): R. G. Bashforth, J. E. Hurst, W. M. Larke, B.A., E. Marks, M.Sc.(Tech.), Ph.D., A. W. Shore, K. H. Wright.

Manchester Metallurgical Society

President: J. D. Hannah, M.Sc.(Tech.)

Vice-Presidents: C. J. Bushrod, M.Sc., G. A. Cottell, M.Sc., K. M. Entwistle, M.Sc., Ph.D.

Hon. Secretary: E. Holland, B.Sc.

Hon. Treasurer: N. Youatt

Members of Council: V. B. Hysel, M.Sc., J. Y. Davies, B.Sc., W. Ashelby, C. F. Brereton, A. B. Ashton, M.Sc., E. F. Emley, Ph.D., F. G. Haynes, M.Eng., W. A. Jenks, W. Sorby, B.Met.

North Wales Metallurgical Society

Chairman: E. Taylor-Austin

Vice-Chairman: R. L. Willott

Hon. Treasurer: S. Andrew

Hon. Secretary: S. Weir

Assistant Secretary: J. B. Edwards

Press Secretary: P. R. Higginson

Committee: C. H. Bacon, H. Edwards, M. Hall, H. H. Johnson, L. Leech, G. Millington, A. D. Osborne, J. Till.

INSTITUTE OF METALS

Mechanism of Phase Transformations in Metals

The Metal Physics Committee of the Institute of Metals is organizing a one-day Symposium on "*The Mechanism of Phase Transformations in Metals*," to be held probably during November, 1955. The Symposium will be in two parts: Transformations that are largely governed by thermal activation (nucleation and growth reactions), and transformations in which thermal activation is less important (martensitic reactions).

Review papers will be invited to introduce each session, but in addition the Committee will be glad to consider offers of short original contributions (up to 3000 words) from workers in the field of phase transformations. In these contributions, emphasis should be laid on the experimental results and their interpretation, rather than on details of experimental techniques.

EDUCATION

Joint Committee on Metallurgical Education

By mutual agreement between The Iron and Steel Institute, the Institution of Mining and Metallurgy, The Institute of Metals, the Institute of British Foundrymen, and The Institution of Metallurgists, it has been decided that the Joint Committee on Metallurgical Education (Chairman, Mr. D. R. O. Thomas) should be dissolved at the end of 1954 and that future work within the field of the Committee's interests should be carried on by The Institution of Metallurgists through the Institution's Education Committee.

In order to maintain continuity (but to avoid duplicating Committees) and to provide for the continued interest in metallurgical education of the other societies, The Institution of Metallurgists has agreed to invite each of the other four societies mentioned above to appoint two representatives to join the Institution's Education Committee. Representation of other interests hitherto represented on the Joint Committee will be considered later.

The Iron and Steel Institute has provided secretarial services for the Joint Committee on Metallurgical Education, since it was formed in 1945. A note on the work of this Committee will be published in a later issue of the *Journal*.

Joint Committee on National Certificates in Metallurgy

By agreement between The Iron and Steel Institute, the Institution of Mining and Metallurgy, The Institute of Metals, and The Institution of Metallurgists, and with the concurrence of the Ministry of Education, secretarial services for the Joint Committee for National Certificates in Metallurgy are being provided by The Institution of Metallurgists instead of by The Iron and Steel Institute from 1st January, 1955. Accordingly, Dr. A. D. Merriman, Registrar-Secretary of The Institution of Metallurgists, will serve as Secretary of the Joint Committee in place of Mr. K. Headlam-Morley, who has been Secretary since the Committee was formed in 1945. By mutual arrangement, Miss M. E. Thompson, who has assisted the Secretary in carrying out the work of the Joint Committee, has left the staff of The Iron and Steel Institute and has joined the staff of The Institution of Metallurgists.

All communications should in future be addressed to:
The Secretary,
Joint Committee for National Certificates
in Metallurgy,
c/o The Institution of Metallurgists,
28 Victoria Street,
London, S.W.1.

Report on 1953–1954

Courses approved by the Joint Committee were in operation during the session 1953–1954 as follows: Senior Courses for the Ordinary National Certificate (37), Contributory Centres, with Courses for the First Year or First and Second Years of the Senior Course (6), Advanced Courses for the Higher National Certificate (20).

258 candidates have qualified for the award of an Ordinary National Certificate in Metallurgy, and 144 for the award of a Higher National Certificate. Distinctions have been gained by 25 candidates for the Ordinary Certificate and 22 for the Higher. Prizes were awarded to 39 successful candidates for the Ordinary Certificate and 31 for the Higher.

Comparative figures of entries and successful candidates for the nine years since the institution of the Scheme for National Certificates in Metallurgy in England and Wales are as follows:

Ordinary National Certificates

Year	Technical Colleges	Entered	Number of Candidates Awarded Certificates	With Distinction	With Prize Award
1946	4	28	23	4	4
1947	11	102	69	14	17
1948	15	123	74	12	16
1949	21	222	107	24	27
1950	20	281	142	23	26
1951	25	314	138	17	24
1952	28	334	186*	22	31
1953	31	390	184	17	26
1954	37	522	258	25	39
		2316	1181		

Higher National Certificate

Year	Technical Colleges	Entered	Number of Candidates Awarded Certificates	With Distinction	With Prize Award
1946	—	—	—	—	—
1947	2	22	18	—	1
1948	8	38	22	1	2
1949	10	74	59	20	22
1950	11	62	42	8	10
1951	16	110	54	5	11
1952	16	132	120*	16	22
1953	17	140	116	17	24
1954	18	173	144	22	31
		751	575		

Assessed Examinations in Mathematics S.2 and S.3

Year	Technical Colleges	Candidates Entered	Passed Examinations
1952	5	82	46
1953	11	176	129
1954	22	341	238
		599	413

The Joint Committee has noted with pleasure the increase in the number of candidates for the Ordinary and Higher Certificate in 1954.

Attention was drawn in the Joint Committee's Reports for 1952 and 1953 to the regulations for admission to the examinations of The Institution of

* The results of 8 candidates for the Ordinary Certificate and 17 candidates for the Higher Certificate whose records were not available in 1951 were included in 1952.

Metallurgists. Candidates holding an Ordinary National Certificate in Metallurgy in which Chemistry and Physics are assessed subjects in the final year and Mathematics an assessed subject in the pre-final year, will be regarded as satisfying the regulations for admission to the Institution's examinations for Licentiateship. The regulations also provide for partial exemption from the Licentiateship examination to holders of the Higher National Certificate, provided that the standard attained in a minimum of three subjects approaches the standard required for distinction. The Joint Committee is pleased to report that 15 of the successful candidates for the Higher National Certificate in Metallurgy in 1952, 11 in 1953, and 21 in 1954 attained the exemption standard of The Institution of Metallurgists and qualified for exemption from the Licentiateship examinations in three or more subjects.

Mond Nickel Fellowships

The Mond Nickel Fellowships Committee announces the following awards for 1954:

M. Brownlee (Dorman, Long and Co., Ltd., Redcar), to study British, Continental, and American hot and cold metal basic open-hearth steelmaking, with particular reference to furnace design and construction and factors affecting ingot quality.

R. D. Butler (Imperial Chemical Industries, Ltd., Liverpool), to study mineral dressing practice in Great Britain, on the Continent, and in the U.S.A. and Canada, with particular reference to the design, layout, and operation of small-scale mills.

F. B. Peacock (Dorman, Long and Co., Ltd., Middlesbrough), to study rolling-mill operation and maintenance with particular reference to soaking pits, blooming mills, and the production of universal beams, structural sections, and rails.

A. M. Sage (British Iron and Steel Research Association, London), to study the manufacture and fabrication of structural steels in the United Kingdom, Europe, U.S.A., and Canada.

NEWS OF SCIENCE AND INDUSTRY

National Association of Corrosion Engineers

The Eleventh Annual Conference and Exhibition of the National Association of Corrosion Engineers will be held on 7th–11th March, 1955, at Palmer House, Chicago, Illinois, U.S.A.

Institute of Physics

The 1955 Summer Meeting of the Non-Destructive Testing Group of the Institute of Physics will be held in the Societies' Room of the Royal Technical College, Glasgow, on 22nd–25th June.

Protection of Structural Steel

A Symposium on "*The Protection of Structural Steel*," organized by the Corrosion Group of the Society of Chemical Industry, will be held in the lecture-theatre of the Institution of Civil Engineers on 31st March and 1st April, 1955. The papers to be presented will include accounts of practical experience and of research with protective coatings from several countries. Contributions have been promised from the U.S.A., France, Belgium, Holland, and Sweden, and British contributors will include W. E. Ballard and F. A. Rivett, F. Fancutt, F. R. Himsworth, J. C. Hudson and J. F. Stanners, L. A. Ravald and W. A. Johnson.

Non-members of the Corrosion Group who wish to receive a copy of the full notice of meeting and a registration form, when it becomes available, should apply to the Hon. Secretary of the Group, Mr. S. C. Britton, Tin Research Institute, Fraser Road, Greenford, Middlesex.

Determination of Carbon in Ferrous Metals

A new method for the determination of very low carbon contents in certain ferrous materials used for electrical purposes has been developed in the Research Laboratories of the General Electric Co., Ltd. The method involves the electrical conductimetric determination of the carbon dioxide produced in the combustion of a sample of normal size (1 or 2 grammes).

Data Sheets on High-Nickel Alloys

A new series of data sheets giving essential physical and mechanical data on wrought high-nickel alloys is being issued by Henry Wiggin and Co., Ltd. The first in the series deals with the use of Nimonic 90 for springs operating at extreme temperatures; the second gives data on Monel, a corrosion-resisting nickel–copper alloy of wide engineering application. Copies of both these data sheets are obtainable on request to the Publications Department, Henry Wiggin and Co., Ltd., Thames House, Millbank, London, S.W.1.

British Cast Iron Research Association

The British Cast Iron Research Association has recently issued its 33rd Annual Report for the year 1953–54. The Report gives details of the work of the Research Department during the year on such subjects as dephosphorization, corrosion, cracking, mechanical properties, solidification of cast irons, and methods of analysis; the work of the Development Department in providing a liaison between research and industry is also covered. The Operational Research Team has also been active and has made 125 visits to member-firms of the Association during the year.

Nuclear Energy

Developments in the use of nuclear energy, especially in the construction of nuclear power stations, have now reached a stage where a considerable number of research projects are in hand in many countries. To provide a channel for the communication of ideas on this subject Pergamon Press have decided to publish a new periodical "*Journal of Nuclear Energy*," to be issued quarterly at first. The first number has now appeared.

Coke Oven Plants

Woodall-Duckham Construction Co., Ltd., have received from the South African Iron and Steel Industrial Corporation Ltd. two large orders for coke oven plants. One plant is to be built at the Vanderbijl Park Works and the other at the Pretoria Works. The total value of the two orders is approximately £1¾ million.

Loewy Engineering Company

The new drawing and design offices of the Loewy Engineering Co., Ltd., Wallisdown Road, Bournemouth, were opened on 7th December, 1954, by Sir Edward C. G. Boyle, Parliamentary Secretary to the Ministry of Supply.

Changes of Name and Address

The Northern Area office of Birlec Ltd. has removed to 317 Glossop Road, Sheffield, 10 (Tel. Sheffield 63258-9).

The Water Pollution Research Laboratory of the Department of Scientific and Industrial Research has removed to Elder Way, London Road, Stevenage, Herts. (Tel. Stevenage 820).

Koninklijke Nederlandsche Staalfabrieken v/h J. M. de Muinck Keizer N.V., of Utrecht, Holland, will in future be known as Koninklijke Demka Staalfabrieken N.V.

The Manchester and North Midlands Branch Office of the JOHN THOMPSON GROUP of Companies has removed to new premises at Sunlight House (6th Floor), Quay Street, Manchester, 3 (Tel. Deansgate 6151–2).

NORMAN W. HUTCHINSON AND SONS PTY., LTD., of Melbourne, Australia, have incorporated a London office under the name of Norman W. Hutchinson and Sons (U.K.) Ltd., at Hanover Court, Hanover Street, London, W.1.

The Bedford Depot of BRITISH INSULATED CALLENDER'S CABLES LTD. has been closed, and a new Depot has been opened at 81 Dumfries Street, Luton (Tel. Luton 6866).

The Middlesbrough District Office of BRITISH INSULATED CALLENDER'S CONSTRUCTION CO., LTD., is now situated at 55–57 Borough Road, Middlesbrough (Tel. Middlesbrough 43644).

THE FLAME RADIATION RESEARCH BRITISH COMMITTEE has removed to new offices at 6 De Vere Gardens, London, W.8 (Tel.: Western 0761).

Industrial Publications Received

▶ Fully illustrated brochure on "*Industrial Stoving Equipment*"—Publication No. 0/1—A. E. Griffiths (Smethwick) Ltd.

▶ The International Furnace Co., Ltd., has published an illustrated booklet describing a two-stage gas producer plant. Two plants of this kind are already in operation, one producing heating gas for steelworks furnaces with electrostatic gas cleaning at the River Don works of the English Steel Corporation Ltd., and the other producing gas for tube manufacture with electrostatic gas cleaning at the Wednesfield works of the Weldless Steel Tube Co., Ltd. The daily throughput of the plant at the River Don works is 120 tons of pit coal; there are two units, and the net calorific value is between 1500 and 1600 cal.

▶ "*Three-Metre Grating Polychromator with Direct Photoelectric Recording of Spectrum Line Density*"—Publication No. CH.385—"*Accessories for Infra Red Spectrophotometry*"—Publication No. CH.388/R—Hilger and Watts Ltd.

▶ Brochure on boiler-house coal and ash handling plant—Marco Conveyor and Engineering Co., Ltd.

▶ The Model '72' freightlifter, a 5-ton capacity fork-lift truck, is described in a brochure issued by Shelvoke and Drewry Ltd.—Publication No. FLT 7/54.

DIARY

3rd Feb.—INSTITUTE OF METALS (London Local Section)—"*Tin and Tinplate*," by W. E. Hoare—4 Grosvenor Gardens, London, S.W.1, 6.30 P.M.

3rd Feb.—INSTITUTE OF METALS (Birmingham Local Section)—"*The Effect of Trace Elements on the Properties of Iron and Steel*," by N. P. Allen—James Watt Memorial Institute, Great Charles Street, Birmingham, 6.30 P.M.

3rd Feb.—STAFFORDSHIRE IRON AND STEEL INSTITUTE—"*Work Study Applied to Steel Production*," by T. U. Matthew—Star and Garter Royal Hotel, Wolverhampton, 7.30 P.M.

7th Feb.—CLEVELAND INSTITUTION OF ENGINEERS—"*Iron Making at Skinningrove*," by R. Kingston—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

8th Feb.—INSTITUTE OF METALS (South Wales Local Section)—"*Tool and Die Steels*," by E. Johnson—Metallurgy Department, University College, Singleton Park, Swansea, 6.45 P.M.

8th Feb.—SHEFFIELD METALLURGICAL ASSOCIATION—Discussion on Open-Hearth Furnaces: Acid and Basic Bottoms—B.I.S.R.A., Hoyle Street, Sheffield, 3, 7.0 P.M.

9th Feb.—NORTH WALES METALLURGICAL SOCIETY—"*Enamelling of Iron*," by A. Biddulph—County Primary School, Plymouth Street, Shotton, Chester, 7.15 P.M.

9th Feb.—MANCHESTER METALLURGICAL SOCIETY—"*Industrial Applications of Precious Metals*," by J. C. Chaston—Lecture Room, Central Library, Manchester, 6.30 P.M.

10th Feb.—EAST MIDLANDS METALLURGICAL SOCIETY—Members' Night—Nottingham and District Technical College, Shakespeare Street, Nottingham, 7.30 P.M.

10th Feb.—EBBW VALE METALLURGICAL SOCIETY (Joint meeting with the Engineers Group)—"*The Design and Use of Some Rolling Mill Instruments and Controls*," by R. B. Sims—Lecture Hall, Richard Thomas and Baldwins Ltd., Ebbw Vale, 7.15 P.M.

10th Feb. — LIVERPOOL METALLURGICAL SOCIETY — "*Thermochemistry of Alloys*," by A. R. Harding—Liverpool Engineering Society, The Temple, Dale Street, Liverpool, 7.0 P.M.

12th Feb.—SWANSEA AND DISTRICT METALLURGICAL SOCIETY—"*The Initial Organisation and Operation of the Abbey Rolling Mills*," by H. H. Ascough—Central Library, Swansea, 6.30 P.M.

14th Feb.—LINCOLNSHIRE IRON AND STEEL INSTITUTE—Young Members' Meeting—North Lindsey Technical College, Kingsway, Scunthorpe, 7.30 P.M.

15th Feb.—SHEFFIELD METALLURGICAL ASSOCIATION (Joint meeting with the Royal Institute of Chemistry)—"*Titanium*," by N. P. Inglis—B.I.S.R.A., Hoyle Street, Sheffield, 3, 7.0 P.M.

17th Feb.—STAFFORDSHIRE IRON AND STEEL INSTITUTE (Associates' Section)—Film evening—Wolverhampton and Staffordshire Technical College.

18th Feb.—MANCHESTER ASSOCIATION OF ENGINEERS—"*Some Aspects of Drop Forging Practice*," by K. Fidler—Engineers' Club, Albert Square, Manchester, 2, 6.45 P.M.

21st Feb.—CLEVELAND INSTITUTION OF ENGINEERS—"*Storage, Handling, and Usage of Oil Fuel in the Steel Making Industry*," by H. W. Hastings—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

22nd Feb.—NORTH EAST METALLURGICAL SOCIETY—"*Plain Bearing Metals*," by P. T. Holligan—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 7.15 P.M.

23rd Feb.—MANCHESTER METALLURGICAL SOCIETY—"*Brittle Fracture*," by T. S. Robertson—Lecture Room, Central Library, Manchester, 6.30 P.M.

1st Mar.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Segregation in High-Speed and Other Tool Steels*," by E. Johnson—B.I.S.R.A., Hoyle Street, Sheffield, 3, 7.0 P.M.

3rd Mar.—INSTITUTE OF METALS (Birmingham Local Section)—"*Recent Developments in Stainless Steels*," by J. I. Morley—James Watt Memorial Institute, Great Charles Street, Birmingham, 6.30 P.M.

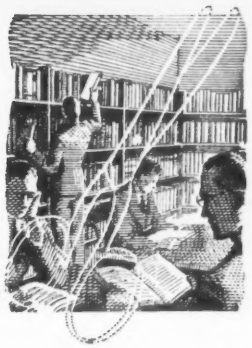
TRANSLATION SERVICE

(The previous announcement was made in the January, 1955, issue of the *Journal*, p. 78).

TRANSLATION AVAILABLE

No. 498 (Swedish). C. G. CARLSSON and L. DANIELSSON: "Direct-Reading Spectrographic Determination of Phosphorus in Steel and Iron Ore." (*Jernkontorets Annaler*, 1954, vol. 138, No. 7, pp. 383–403).

ABSTRACTS OF CURRENT LITERATURE *and* BOOK NOTICES



IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

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MINERAL RESOURCES

Newfoundland. A. Buisson. (*Min. Mag.*, 1954, **91**, Aug., 82–91). An account is given of the history, resources, and development of Newfoundland and Labrador. In Newfoundland the three mines of the Wabana group (Dominion Steel and Coal Corp.) produce hematite, the reserves being estimated at 4×10^9 tons; new developments are described. In Labrador, large deposits of high-grade ore are being exploited along the Quebec boundary by the Iron Ore Co. of Canada, and exploration has disclosed further substantial deposits of material probably capable of concentration.—K. E. J.

Mineral Resources of the Region between the Vosges and Ardennes. V. Charrin. (*Génie Civil*, 1954, **131**, May 15, 188–192). The mineral deposits of this region in which all geological epochs are represented are described. The iron ores are high in phosphorus but suitable for the basic-Bessemer process. Reserves of Minette ore total 4,830 million tons and these contain appreciable quantities of titanium and vanadium.—A. G.

Magnesite Production in Yugoslavia. (*Min. J.*, 1954, **242**, June 18, 738). A short account of the geological formation of the magnesite deposits is given and the methods used for the processing of the raw material are described.—B. G. B.

The Mineral Resources in Finland. H. Stigzelius. (*Teknillisen Kemian Aikakausilehti*, 1954, May 15, 267–372). [In Swedish]. In this review it is shown that there is no coal to be mined in Finland, but there is an abundance of peat. Iron pyrites and magnetite are available, and the country is self-supporting as regards zinc, gold, and sulphur.—R. A. R.

Mineral and Fuel Resources. R. J. Lund and D. D. Moore. (*Indust. Eng. Chem.*, 1954, **46**, Mar., 483–487). Resources of the five Eastern North-Central states of the U.S.A. are reviewed. Future supplies of coal, limestone, dolomite, iron, silica, and fluospar are assured. By-products of industries such as that of iron and steel will provide large amounts of chemical raw materials.—A. G.

ORES—MINING AND TREATMENT

Technical Equipment in Sampling Stations. J. Freygang. (*Stahl u. Eisen*, 1954, **74**, July 15, 948–951). The various types of equipment (grinding mills, drilling and milling machines, sieves, drying ovens, weighing machines, etc.) which are used in sampling stations are described and illustrated.—J. P.

Mechanical Trimmers for Ships Carrying Iron Ore. H. R. Mills and J. G. Lee. (*Mech. Handling*, 1954, **41**, June, 350–353; Aug., 475–481).

The Significance of Agglomeration in the Mineral Industries. E. Swartzman. (*Canad. Min. Met. Bull.*, 1954, **47**, May, 318–327). After a general introduction it is pointed out that the terms in use have become confused, and a standard terminology is suggested. Processes defined are compacting, pelletizing or balling, and sintering. Suggested terminology for products includes compact, granule, pill, tablet, briquette, brick, preform, spheroid, ball, pellet and pellet granule. Various types of agglomerating machines are described and discussed. Compacting presses are either of the closed mould variety, which includes plunger presses (single or double action), roll-presses, ring-roll presses, and tamping presses, or of the open-mould type, which covers horizontal extruders, vertical extruders, and ring-roll extruders. Pelletizing by tumbling is divided into dry and wet pelletizing; there are two main wet processes, employing either a single horizontal rotating drum, or two rotating drums in parallel, and also a trough pelletizer and a pan pelletizer. Finally, sintering is mentioned. Some details of size and capacities of plant are given. (19 references).—T. E. D.

Burden Preparation at Société Métallurgique de Knutange. Thierry. (*Centre. Doc. Sid., Circ. Inform. Tech.*, 1954, (2), 339–350). A description is given of the crushing plant and Dwight Lloyd sintering plant at Knutange. Results for furnaces operating on a raw burden and with 55% sinter show that control is more critical with the latter. Given good management, the use of sinter results in better coke utilization and superior iron.—A. G.

Ore Dressing Research in India—1953. (*Chemical and Process Engineering*, 1954, **35**, Apr., 113–114). The recent developments in India in the organized study of the principles and practice of ore dressing are outlined. Contact angle measurements, flotation, and heavy-media separation studies are in progress on low-grade ores, and the beneficiation of some lean chrome ores has been completed. (15 references).—L. E. W.

Operational Results with the Sink and Float Process for Iron Ore Dressing. A. Goltz, W. Jacobs, and O. Burghardt. (*Stahl u. Eisen*, 1954, **74**, Aug. 12, 1070–1075). The heavy-medium equipment at the Erzbergbau Salzgitter A.G., Calbecht, has fulfilled economic and technical expectations. This flotation process fills a gap left by previous dressing methods. The plant at the Morgenstern mine of the Barbara Erzbergbau A.G. has proved its worth after overcoming initial difficulties, and has led to a marked improvement in the Fe/SiO_2 ratio in Salzgitter ores. Operational results are tabulated. The satisfactory results at the Alte Hütte, Wissen, Siegerland, indicate that the sink and float process will be suitable for dealing with ores from other local mines.

FUEL—PREPARATION, PROPERTIES, AND USES

International Flame Radiation Research. (*Metallurgia*, 1954, **49**, June, 265–270: *Refract. J.*, 1954, **30**, May, 212–215: *Colliery Guardian*, 1954, **188**, June 24, 781–785: *Iron Steel*, 1954, **27**, June, 229–231: *Usine Novelle*, 1954, **10**, May 20, 29–31). A description of the new experimental station which has been constructed at Ijmuiden in Holland for international research on flame radiation, is given. Two furnaces well insulated and able to withstand temperatures of 1600° and 1500° C. have been specially designed for the investigation of different kinds of flame.

Modern Preparation Plant Assures Clean Coal of Uniform Size. A. L. Lynn and J. H. Green. (*Blast Furn. Steel Plant*, 1954, **42**, July, 787–790). A short account of a new coal cleaning plant installed by the Island Creek Coal Co., U.S.A., is given.—B. G. B.

Evaluation of the Performance of Coking Plants and Blast-Furnaces in the First Czechoslovak Five-Year Plan. L. Sajch and F. Váňa. (*Hutnické Listy*, 1954, **9**, (6), 321–327). [In Czech]. A comparison is made of the annual productions and corresponding qualities and yields of coke over the period 1949–53. Histograms for pig-iron production, blast-furnace charge analyses, and changes in operating conditions in the blast-furnaces of four major steelmaking plants over the same period are given. On the basis of the analyses proposals for improvements in the production of coke and iron are advanced.

The Manufacture and Use of Ferro-Coke. (*Génie Civil*, 1954, **131**, June 1, 212–213). Preliminary trials have shown that coke can be made from fine ore and gas coals of high volatile content provided carbonization is sufficiently rapid. The oxygen in the ore increases the gas volume but lowers its calorific value. In industrial trials 6000 tons of ferro-coke have been made from gas coal with 30% magnetite concentrates without any practical difficulties. Depending on its size, ferro-coke may be used in the blast-furnace, the low-shaft furnace or after preliminary agglomeration.—A. G.

Coke Particle Size. E. Ferrara. (*Calore*, 1954, **25**, June, 254–263). [In Italian]. The author discusses the coke sizing, gives definitions, reviews Italian, British and German nomenclature and practice, and examines the differences which exist between the laboratory and the industrial approach to the subject.—M. D. J. B.

Gasification of Solid Fuel: OEEC Report on Technical Progress in the U.S.A. and France. (*Iron Coal Trades Rev.*, 1954, **169**, July 16, 137–141). A summary is given of part of an O.E.E.C. report dealing with work done during the last few years on gasification of solid fuels under pressure, the main features of some of the more promising new processes being indicated. These include the Flesch-Demag, Lurgi-Ruhrgas, Panindco, and Imperial Chemical Industries moving-bed systems, and descriptions are also given of American developments, including the Koppers-Totzek producer, and of work on the Vortex reaction chamber at the Fuel Research Station, Greenwich.—G. F.

A New Approach to the Theory and Operation of Electrostatic Precipitators for Use on Pulverized-Fuel-Fired Boilers. N. Troost. (*Proc. Inst. Elec. Eng.*, 1954, **101**, Part II, Aug., 369–389).

Gas and the Steel Industry. E. J. Pode. (*Brit. Steelmaker*, 1954, **20**, June, 198–199). The author discusses the considerable joint interests of the gas industry and the iron and steel industry, dealing mainly with completely integrated composite iron and steel plant.—G. F.

Steel and Oil. (*Brit. Iron Steel Monthly Stat. Bull.*, 1954, **29**, July, 5–8). The use of heavy fuel oil as an alternative to coal in the British steel industry is discussed.—V. G.

Stability and Suitability of Fuel Oils. B. Riediger. (*BWK*, 1954, **6**, Aug., 307–312). Economic problems involved in the storage, and mixing of various types are analysed, and tests for storability, such as cloud and pour-point determinations, their significance and adequacy are discussed. The chemical structure of common fuel oils, their solubility in various solvents, the processes occurring in mixing and during storage, and problems of assessing storability and suitability are considered.—P. F.

Fuel Oil Measurements with a Transmitter. F. Strähuber and A. Richter. (*Stahl u. Eisen*, 1954, **74**, July 15, 960–961). A compressed air operated transmitter is described. This is for the remote indication of the pressure difference across a diaphragm, jet, or venturi tube in an oil pipeline, for oil flow measurement.

TEMPERATURE MEASUREMENT AND CONTROL

The Disappearing Filament Optical Pyrometer. A. E. Turner. (*Instruments in Industry*, 1954, **1**, Aug., 73–78). The construction and properties of high precision laboratory and industrial disappearing filament optical pyrometers are described and compared. Details are given of operation, necessary corrections, and maintenance.—L. D. H.

Some Consideration of the Errors of Brightness and Two-Colour Types of Spectral Radiation Pyrometer. E. C. Pyatt. (*Brit. J. Appl. Phys.*, 1954, **5**, July, 264–268). The limitations and derived errors of the two types of pyrometer are considered and the construction of comprehensive correction charts described. The practical application of a graph of ϵ_1 *versus* ϵ_2 (emissivities at the two-colour pyrometer wavelengths λ_1 and λ_2 corresponding to a given ratio Z between the errors of the two-colour and brightness pyrometers, the latter working at λ_2) is demonstrated and the possibility of extended use of this relationship with a more complex ratio pyrometer is discussed.—J. O. L.

The Theoretical Characteristics of Bichromatic Pyrometers. H. Herne. (*Brit. J. Appl. Phys.*, 1954, **4**, Dec., 374–378). The theory of the bichromate pyrometer is outlined and its response is calculated for different filter pass bands of different band widths. The ratio of the response at one colour to the response at the other is calculated as a function of the temperature of the source and the results are presented graphically.—R. A. R.

Design and Construction of Needle Thermocouples. W. G. Rauch. (*Metal Progress*, 1954, **65**, Mar., 71–74). Thermocouples of very small cross-section have been constructed from hypodermic tubing and a central wire; the central wire being insulated from the tube. Details of the construction of such couples, suitable metals and alloys, and their performance are given. Couples consisting of an insulated constantan wire in an Inconel tube have been used successfully up to 1110° F.—B. G. B.

New Standards for Electric Temperature Measuring Instruments. H. Lotz. (*Elektrowärme Techn.*, 1954, **5**, May, 100–102). The revised and new German standard specifications relating to thermocouples, electric resistance thermometers, and the general principles of thermometry are discussed.

The Choice of Temperature Recorders. H. Bönnhoff. (*Elektrowärme Techn.*, 1954, **5**, May, 97–99). The choice of modern temperature recorders of the potentiometer, photographic, and photo-cell types is considered.

Temperature Control of Vitreous Enamelling Furnaces. (*Indust. Gas*, 1954, **17**, June, 288–290). A temperature controller designed for use with vitreous enamelling furnaces and activated by a Brown Radiation Detector, a type of total radiation pyrometer, is described.—D. H.

REFRACTORY MATERIALS

The Properties of Fireclay, Silica and Siliceous Refractories in Relation to their Industrial Uses. A. R. Myhill. (*Mech. World*, 1954, **134**, June, 273–276). The characteristics of the more widely used types of refractory brickwork are discussed in relation to their particular uses.—D. H.

Carbon Refractories in Metallurgy. R. Maire. (*Silicates Indust.*, 1954, **19**, Mar., 108–118). The history of the use of carbon as a refractory material is traced with special emphasis on the developments in France. The advantages of the use of carbon bricks in blast-furnaces, the types of bricks and fillers, and their composition and manufacture are discussed.

The Problem of Ladle Lining Refractories in Steel Pouring. L. Halm. (*Silicates Indust.*, 1954, **19**, May, 189–206). The properties and life of ladle linings in use in Europe and America are considered, and the literature is reviewed. Data obtained in the author's experiments on ladle linings and allied refractories used in steel pouring are given and their significance is discussed.—P. F.

Refractory Linings of Siliceous Puddled Clay—Improvements in, and Control of, Ramming and Drying. P. Blanchard. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–34). [In French]. Improved methods for ramming and drying siliceous puddled clay linings in a 2-ton steel melting rotary furnace are discussed.—B. C. W.

Wastage of Refractories by Slag Attack. D. Dixon. (*Chemical and Process Engineering*, 1954, **35**, Apr., 121–123). The author discusses in relation to high alumina refractories the

causes and effects of the various forms of attack which may occur simultaneously. The forms of wastage described are: (*a*) Cracking and shrinkage in service as a result of underkilning or high service temperature; (*b*) solution and erosion by fluid slag; (*c*) thermal spalling from rapid heating; (*d*) thermal spalling from rapid heating and also rapid cooling; and (*e*) softening and deformation by high temperature.

The Migration of Slag Components into Refractory Bricks. K. Konopicky. (*Stahl u. Eisen*, 1954, **74**, July 15, 943–947). Published reports on the behaviour of refractory bricks, as received or contaminated with various oxides, when subjected to a temperature gradient, are reviewed. A general picture is presented of the rate and extent of penetration of various slag components. This picture indicates that, when testing refractories for slag resistance, it is important to take into account the temperature gradient, pore structure, texture, and surface forces in the brick, and the nature of the fluxing agent.—J. P.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

A Note on Burden Preparation for Lorraine Blast-Furnaces. Thibaut. (*Centre. Doc. Sid. Circ. Inform. Tech.*, 1954, (2), 327–329). Present practice in the preparation of ore and coke is described. Data are presented and discussed for 44 operational periods of blast-furnaces making basic-Bessemer iron with run of mine ore, ore with a high proportion of fines, crushed and sintered ore, and crushed and screened ore.

The First Operational Results on the Low-Shaft Furnace at Ougrée Marihaye. (*Usine Nouvelle*, 1954, **10**, Apr. 8, 65). A brief summary of the origin and aims of the low-shaft furnace project is given. Early results show a smaller working zone than predicted, which will permit a reduction in the height of future furnaces. With 30% oxygen in the blast a product low in carbon and high in sulphur is obtained, but greater carburization is achieved if fine coke is charged. The dust losses were very high but the results are considered to be encouraging.

Factory Experiments of Molten Pig Sampling. M. Saheki and T. Yonezawa. (*Tetsu to Hagane*, 1954, **40**, Feb., 81–87). [In Japanese]. At the blast-furnace plant of Kamaishi works, there was no significant difference between samples at the runner and at the casting machine. When tapping foundry iron, the silicon content rose to a maximum, then fell towards the end; with open-hearth iron, the sulphur content remained constant at first but dropped during the latter third of the tapping.—K. E. J.

The Formation and Decomposition of Hercynite. B. G. Baldwin. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 142–146). [This issue].

Constitution of Blast-Furnace Slags—Viscosity Measurements and the Mechanism of Desulphurization. Kozakevitch. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (2), 354–356). In basic-Bessemer iron, making the CaO/SiO_2 ratio of 1·2 to 1·4 corresponds to a minimum in the viscosity curve. The higher viscosities of other slags are attributed to the commencement of crystallization. Desulphurization of iron by blast-furnace slags is attributed to the formation of a slag–iron emulsion which separates as the removal of sulphur raises the surface tension of the iron.—A. G.

Treatment of Pig Iron after Tapping to Recover Vanadium. F. Houdek and M. Oppelt. (*Hutnické Listy*, 1954, **9**, (6), 327–333). [In Czech]. A description is given of a series of works experiments in which oxygen or oxygen-enriched air was blown into the iron in the runner on its way from the blast-furnace, to obtain a vanadium-rich slag. Better results were obtained by blowing oxygen or air under the surface of the iron in the tundish. The relation between the residual vanadium content, after the removal of the vanadium-rich slag, and the content of silicon and manganese is shown.

The Design History of the Blast-Furnace Tuyere. L. V. Dillon. (*Blast Furn. Steel Plant*, 1954, **42**, July, 800–804). A short account of the development in the design of blast-furnace tuyeres is given and a number of causes of tuyere failure are considered.—B. G. B.

Engineering Devices for Blast-Furnace and Converter Treatments of Ore. C. C. Downie. (*Min. J.*, 1954, **243**, July 16, 70–71). A very brief account is given of some automatic instruments and new equipment which have been used on blast-furnaces and cupolas. These include automatic taphole plugging and signalling devices, and the continuous recording of flue gas composition.—B. G. B.

An Industrial Installation for the Precipitation of Iron Oxide Fumes. P. Leroy, H. Ruelland, and M. Cassier. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 119–125). To cope with the increasing fume problem associated with oxygen refining in steelworks a two-stage process has been developed comprising a primary dry cleaner and secondary wet washer. The plant is described; it gives almost complete dust recovery. The fume is a mixture of iron oxides which settles rapidly in water and has a commercial value.—A. G.

Effect of Physical and Chemical Properties of Coke on Its Combustion. R. A. Mott. (*Coke Gas*, 1954, **16**, May, 189–194). The combustion of coke in the blast-furnace is discussed with particular reference to recent work carried out at the Midland Coke Research Station. The aspects considered include coke size and porosity, the voidage of coke beds, the resistance of coke beds to the passage of gas, and the production of breeze.

The Influence of Pressure on the Reduction of Metal Oxides by Carbon. E. Iwanciw. (*Archiwum Gornictwa i Hutnictwa*, 1953, **1**, (2), 131–183). [In Polish]. The mechanism of reduction of nickel oxide and iron ore by carbon was studied. From experimental evidence the author postulates that the reduction by carbon is a complex process and consists of: (1) Gaseous reduction: $MO + CO = M + CO_2$; (2) Boudouard reaction: $CO_2 + C = 2CO_2$; and (3) ' fissure diffusion ' of CO and CO_2 through the film of reduced metal. The latter stage explains why the rate of reduction of metal oxides is much higher than that of the diffusion of carbon or gas in metals. Since the overall reaction rate depends on the slowest reaction, ' fissure diffusion ' is the rate controlling process under industrial conditions. Under chosen laboratory conditions ' fissure diffusion ' may be greatly increased and the Boudouard reaction will then be rate controlling. The effect of pressure is only clearly marked in the range of values under vacuum, as the rate of reduction decreases with decreasing pressure, thus clearly indicating the gaseous character of the reduction process.—V. G.

Explosion in a Cowper Stove. Dardenne. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 127–132). An account of a recent explosion is given with an analysis of the probable causes and a description of the arrangements made to prevent its recurrence.—A. G.

The Norwegian Production of Electric Furnace Pig Iron, and Its Properties. J. Sissener. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–27). [In French]. The development of the electric furnace process in Norway for the production of pig iron is briefly reviewed and the present position summarized. The iron contains 0·10–0·70% vanadium and 0·25–0·5% titanium, and results are presented to show that diesel-engine cylinder jackets cast from it have a much longer service life than those cast from normal irons.—B. C. W.

TREATMENT AND USE OF SLAGS

Mechanism of the Fertilizing Action of Dephosphorizing Slags. G. Barbier and S. Trocmé. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 163–173). The fertilizing action of slag produced by a direct granulation process has been compared with that made by fine grinding after slow cooling. Efficiency is mainly governed by the ability to dissolve and although the action of granulated slag may be slower, particularly in non-acid soils, this is not necessarily a disadvantage.

Slag Wool. Heczko. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 139–146). Over 600,000 tons per year of this material are produced annually in America. Blast-furnace slag is reheated in a cupola and then blown with steam or air at the taphole. Details of the operation are given and the mechanism of formation of fibres is discussed. Finally, the properties and uses of slag wool are outlined.—A. G.

Pumice Slag. Heczko. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 135–137). The types of equipment available for the production of this form of slag are described. The most suitable slags are those from blast-furnaces making foundry iron or metal for the open-hearth furnace. After grading the blown slag makes an excellent constructional and insulating material.—A. G.

Methods for the Production of Slag Sand. P. Grossstück. (*Stahl u. Eisen*, 1954, **74**, July 29, 1011–1015). Methods for granulating blast-furnace slag are described and the properties of this sand are discussed, in particular their hydraulic

properties, as measured by the glass and moisture contents. The water content plays an important part and it is well worth while reducing the moisture in the sand and thereby the water consumption.—J. P.

PRODUCTION OF STEEL

Fairless Works. T. J. Ess. (*Iron Steel Eng.*, 1954, **31**, June, r62–r92). A full description is given of the Fairless Works of United States Steel Corp. Raw material handling facilities, coke plant, blast-furnaces, open-hearth plant, and the slabbing, blooming, and billet mills are described as well as the 10-in. continuous bar mill, the hot strip mill, the sheet and tin mills, and the tube division which consists of a skelp mill, two butt-weld tube mills, and finishing equipment.

New Open Hearth Furnaces at Stelco. E. T. W. Bailey. (*Iron Steel Eng.*, 1954, **31**, June, 91–98). A description is given of the plant expansions of the Steel Co. of Canada which include docks, coke ovens, a blast-furnace, and a new open-hearth shop.—M. D. J. B.

Republic's New Open Hearth Furnaces at Cleveland. E. C. McDonald. (*Iron Steel Eng.*, 1954, **31**, June, 99–106). A description is given of the extensions to the Cleveland works of Republic Steel Corp. These include 126 coke ovens, a 1470 short tons per day blast-furnace with steam generating and blowing equipment, four 275-ton open-hearth furnaces, 21 soaking pits and a 72-in. tandem cold reducing mill.

Oxygen in Steelmaking: Recent British Applications. (*Iron Coal Trades Rev.*, 1954, **168**, June 18, 1479–1481). Following an outline of Continental experience, three recent British developments in the application of oxygen in steelmaking are described. At the Steel Co. of Wales Ltd., oxygen is blown on to the steel bath for decarburization, using a water-cooled gun inserted through the roof, and at Richard Thomas and Baldwins Ltd., a similar process is carried out using an oxygen lance. Both practices result in appreciable savings in time, fuel, and furnace refractories. At Skinningrove Iron Co. Ltd., hot metal is de-siliconized by lancing in the ladle, resulting in reduced metallurgical load, less refractory wear, and hotter metal.—G. F.

The Oxygen Steel Making Process. F. J. McMulkin. (*Trans. Canad. Inst. Min. Met.*, 1954, **57**, 225–229: *Canad. Min. Met. Bull.*, 1954, **47**, June, 381–385). After mentioning the advantages of steel produced by surface blowing the hot metal with oxygen, the theory of the process is discussed. Results of using this refining process at Linz, Donawitz, and the Dominion Foundry and Steel Co. are given. The latter attains a phosphorus content in the finished steel as low as 0·013% by using a higher lime charge than that used in the Austrian plants. Slag compositions and oxygen consumption data are given.—T. E. D.

Design and Arrangement of Demag Lance-Blowing Converter Installations. H. Hofmeister and C. Herrmann. (*Demag News*, 1953, (134), 1–4). In this type of converter a high velocity jet of oxygen is blown from above on to or into the bath of hot metal by means of a water-cooled lance; the converter not being blown from the bottom. The Demag converter has a detachable bottom which facilitates relining. Diagrams and photographs of a lance-blown converter are reproduced.—B. G. B.

Waste Gas Conditions in the Bessemer Steel Plant when Using Different Kinds of Blast. H. Kosmider, H. Neuhaus, and H. Kratzenstein. (*Stahl u. Eisen*, 1954, **74**, Aug. 12, 1045–1053). The total waste gases when using various types of blast have been examined in the Bessemer plant of the Klöckner-Hüttenwerke, Haspe. The volume of gases from the converters and in the stacks has been measured and their compositions determined. The waste gas and air leak volumes did not correspond with recognized values. The sensible heat escaping from the converters in waste gases represents a great heat loss. The dust in the gases and the iron losses were determined. No brown fumes and practically no iron loss were found when blowing with oxygen-steam mixtures. In other cases, the determined values of gas volumes and heat content served as a basis for the erection of gas cleaning and waste-heat recovery plant. It was proved that the evaporation of iron is due to high local temperatures. Reduction of the amount of brown fumes by spraying with water or steam cannot be successful; the fineness of the dust necessitates a special filter installation.—J. P.

Dolomite Linings in a Basic Bessemer Steelworks. M. Recht. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 153–162). The desirable qualities of tar and dolomite are described and practice at two works compared. The results show that differences in lining life must be attributed to factors other than the quality or handling of tar and dolomite.—A. G.

Reactions in the Hearth. Dufraine. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (2), 357). The kinetic movement of coke in front of the tuyeres has been studied in models at the Pompey works and the effect of tuyere design on the zone of turbulence investigated.—A. G.

Heat Balance of a Maerz Furnace at the Marrel Works. Venard. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (2), 361–370). Details of this 80-ton furnace are given. A gas balance and heat balance for the furnace and recuperator system are presented. These revealed that the greatest loss in efficiency was due to air infiltration through the brickwork and elimination of these leaks has resulted in a 10% increase in production and higher exit gas temperatures.—A. G.

Experiences Gained in the Introduction of Soviet Instruments for the Steelmaking Furnaces in the Klement Gottwald Steelworks in Kunchitse. (Moravia). (*Hutnické Listy*, 1954, **9**, (6), 333–341). [In Czech]. The mode of operation and experiences gained with Russian equipment for the automatic control of open-hearth furnaces are discussed, with special reference to the temperature controlling mechanisms. These are not based upon roof pyrometry, but maintain the temperature at levels pre-set by the melter. The principles of of the gas reversal controls and of the instruments for the determination and control of the pressure of the furnace atmosphere are considered.—P. F.

Recovery of Stainless Steel Scrap by Melting in the Basic Open-Hearth Furnace. A. Danihelka. (*Hutnické Listy*, 1954, **9**, (6), 341–347). [In Czech]. New methods of remelting stainless steel scrap are described. Based upon melting in the basic open-hearth furnace, three methods of production are considered. (1) Open-hearth/electric-furnace duplexing without oxygen, for low-carbon sorted scrap. (2) Open-hearth remelting for low-carbon sorted scrap. (3) Open-hearth/electric-furnace duplexing, with oxygen used in the latter. This method is best for mixed or dirty scrap of higher carbon content. The author concludes that production methods utilizing oxygen are less advantageous than (1) or (2), of which (2) is preferred.—P. F.

Automatic Control in the Iron and Steel Industry: Application to Open-Hearth Furnaces and Soaking Pits. S. S. Carlisle. (*Iron Coal Trades Rev.*, 1954, **168**, June 18, 1473–1476). The author describes automatic control systems applied to open-hearth furnaces, soaking pits, and reheating furnaces and shows how the proper application of such systems can result in increased output, safer working, and better products.

Electrochemical Study of the Molten Slags (III). R. Sakagami. (*Tetsu to Hagane*, 1953, **39**, Nov., 1240–1250). [In Japanese]. Using a carbon/silicon-carbide electrode pair, potentials were measured in the CaO–Al_2O_3–SiO_2 and CaO–MgO–Al_2O_3–SiO_2 systems, and the activities of CaO and SiO_2 were calculated. The influences of Al_2O_3 and MgO were indicative of amphoteric behaviour, and it is suggested that, in multiple systems, all except very strongly acidic or basic oxides behave in this way. The CaO/SiO_2 ratio is considered the best index of basicity. (21 references).—K. E. J.

Sulphur Partition between Gas, Slag, and Metal Phases under Steelmaking Conditions. E. T. Turkdogan. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 147–154). [This issue].

Thermodynamics of Carbon Dissolved in Iron Alloys. Part III: Solubility of Carbon in Iron–Sulphur Melts. E. T. Turkdogan and R. A. Hancock. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 155–159). [This issue].

Effect of Si, Mn, P, Al, C, Ni, and Cu on the Mechanism of Sulphur Transfer Across a Slag–Metal Interface. K. M. Goldman, G. Derge, and W. O. Philbrook. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, May, 534–540). Using laboratory melts in induction-heated graphite crucibles, the authors have studied the effects of the above elements on the process of sulphur transfer across a slag–metal interface in a carbon-saturated system. In increasing order of effect, the elements carbon, silicon, manganese, and aluminium all increase the rate of sulphur transfer, whilst the elements copper, nickel, and phosphorus have no observable effect. The reasons for these influences are interpreted in terms of a three-stage mechanism.—G. F.

Economic Comparison of the Open-Hearth and Electric Furnaces for the Manufacture of Unalloyed Low-Carbon Steels. S. L. Case, D. D. Moore, C. E. Sims, and R. S. Lund. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 3–42). A critical survey has been made of the making of unalloyed steels with up to 0·25% carbon by the electric and open-hearth processes in steelworks producing 225,000, 450,000, and 900,000 tons of steel per year, operating on both a solid charge and one incorporating 50% hot metal. The electric process requires a lower capital investment and gives a better return than the open-hearth on an all scrap charge. It is also superior on 50% hot metal at the larger steelworks provided the price of steel remains above 64 dollars/ton. Intangibles such as ease of control and better working conditions must also weigh in favour of electricity.—A. G.

Economics of Melting: The Battelle Electric Furnace Report Reconsidered. J. C. Howard. (*Iron Steel*, 1954, **27**, June 12, 252–255). The author gives further consideration to the Battelle Memorial Institute report comparing the economics of open-hearth and electric furnaces in the production of low-carbon steel. He estimates similar comparative costs for British conditions, showing the economic advantage to be in favour of the electric furnace, and considers that future conditions will favour it even more strongly.—G. F.

The Quality of Toolsteel—Recent Developments and Future Predictions. B. George. (*Metal Progress*, 1954, **65**, Mar., 75–78). Recent developments in the quality of tool steel are considered. Melting in the electric furnace is one of the most important factors in improving the quality. Segregation in ingot moulds and heat-treatment in relation to the production of tool steel are also discussed. The manner in which the quality of tool steel can be improved by testing is considered.—B. G. G.

Determination of the Optimum Current in an Arc Furnace. W. E. Schwabe. (*Iron Steel Eng.*, 1954, **31**, June, 87–90). The author describes the work of a German arc-furnace committee which has studied the problem of practical furnace operation at optimum current and developed a simple mechanism for adjusting the arc current.—M. D. J. B.

Note on the Lining of Electric Furnaces with a Special 'H' Alumino-Silicate Refractory Mix. Hanne. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (2), 371–372). The use of this lining in a 10-ton basic electric furnace has been highly successful. Preparation, ramming, and drying present no special difficulties and partial relining (about once a month) takes only 5–6 hr. An appendix gives operating data for 1951.—A. G.

The Structure of Steelworks Ingot Moulds. M. Ferry. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–36). [In French]. A metallographic study has been made of specimens taken from ingot moulds at various positions through the wall thickness. The structure was ferritic with fine graphite flakes at the mould face, and it showed a gradual transition to a pearlitic structure at the centre of the wall, with an increase in the size of the graphite flakes.—B. C. W.

Principal Results of an Investigation into Ingot Moulds Used in French Basic-Bessemer Steelworks. M. J. Duflot. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–30). [In French]. This is a report on a collaborative investigation carried out in both steelworks and foundries. Among the aspects of the problem which have been considered are: (*a*) The effect of the relative weight of ingot and mould on mould consumption; (*b*) the analyses of ingot mould irons; (*c*) charge compositions and melting practices used in making the irons; (*d*) the effect of ingot mould stripping practice on mould consumption; and (*e*) the causes of ingot mould failure. A statistical analysis is presented of the causes of failure of ingot moulds in 1951 in French steelworks. The best mould life is obtained when the ratio of ingot mould weight to ingot weight is less than 1·2.—B. C. W.

Contribution to the Study of Steelworks Ingot Moulds. J. Durand. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–45). [In French]. Recent work on ingot moulds, particularly that carried out by the Technical Commission on Ingot Moulds formed by Le Centre Technique des Industries de la Fonderie and L'Institut de Recherches de la Sidérurgie, is briefly reviewed.—B. C. W.

Effect of Temperature and Pouring Speed on Ingot Structure. Ingots Committee of B.I.S.R.A. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 120–123). [This issue].

Ingot Heat Conservation. Cooling of 15-ton Ingots between Teeming and Stripping. L. H. W. Savage and M. D. Ashton. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 132–142). [This issue].

Study of a Machine for Descaling Billets by Milling. Charbonnier. (*Centre Doc. Sid. Circ. Inform. Tech.*, 1954, (2), 373–381). A description of the machine and its accessories is given and the factors of control, labour, tool consumption, and economy are considered. The high capital cost of the plant is justified for heavy sections, blooms more than 9 ft. 9 in. long, steels with many and deep surface defects, and for special steels of great value.—A. G.

PRODUCTION OF FERRO-ALLOYS

Sampling of Ferro-Alloys and Their Ores. H. Wirtz. (*Arch. Eisenhüttenwesen*, 1954, **25**, May-June, 225–230: *Iron Steel Inst.*, 1954, *Translation Series No.* 492). Sampling of the more important ferro-alloys and their ores is described. Some criteria for characterizing and assessing the alloys and methods for simplifying sampling are given.—J. P.

Manganese Production by Electrolysis. (*Chemical and Process Engineering*, 1954, **35**, Mar., 89–90). Details are given of the commercial production of pure manganese by electrolysis. Electrolytic manganese has been successfully used in the production of non-ferrous manganese alloys and is particularly suitable for the production of special steels whose characteristics are greatly affected by impurities.

FOUNDRY PRACTICE

Melting and Freezing. B. Chalmers. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, May, 519–532). The author discusses the fundamentals of the processes of melting and freezing, explaining many of the phenomena which have been observed and in particular accounting for the structures that are obtained in actual ingots and castings.

Studies on the Fluidity of Molten Iron (I). Y. Koike. (*Tetsu to Hagane*, 1953, **39**, Sept., 948–953; Oct., 1171–1177). [In Japanese]. Fluidities of molten iron were measured through the flow from a high-frequency furnace through a calibrated magnesia tube. The effects of temperature, composition, and degree of deoxidation were studied. In general, the fluidity increased with increase of temperature and of the content of carbon and silicon. Further observations using a capillary tube method are reported. The fluidity of pig iron was high when its FeO content was high, and the fluidity of 0·3–0·4 carbon steel decreased after deoxidation with aluminium.

The Practical Use of a Ternary Diagram for Alloy Cast Irons. A. W. J. van den Bergh. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–28). [In French]. The author shows how a three dimensional diagram may be used to illustrate the relationships between total carbon, silicon, casting thickness, and structure for cast iron. A similar diagram may be used with alloy irons if the effect of each alloying element on graphitization is expressed as the percentage of silicon that would have the same effect.—B. C. W.

Study of the Flow of Metal in Moulds—Practical Application of Scientific Principles. Experimental Results. C. Trenckle. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–13). [In French]. The filling of a mould and the flow of metal in feeding systems is considered mathematically and equations are derived for calculating the size and spacing of runners and gates. The equations are derived generally, but are applied graphically to castings whose thickness is small relative to their breadth and length. The best results in feeding such castings are achieved with a number of small gates.—B. C. W.

The Viscosity of Liquid Metals Considered as a Phenomenon of Intracrystalline Mutations. E. Piwowarsky and M. Nacken. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–49). [In French]. The viscosity of a liquid metal is considered to arise from the ease or otherwise of atomic movements under stress into vacant sites in the liquid structure, and it is these atomic movements that are referred to as "intracrystalline mutations." From theoretical considerations an equation for the viscosity is derived, and applied to published data on viscosity. The viscosity of metals and alloys is discussed with particular reference to the relevant phase diagrams.—B. C. W.

Equipment for Preheating Blast for the Cupola. I. P. Petrov. (*Liteĭnoe Proizvodstvo*, 1954, (3), 12–14). [In Russian]. An improved design of recuperator for preheating the blast of a cupola with an output of 5 ton/hr. is described. The rate the cupola gas enters the recuperator is kept constant with the aid of a compressed air ejector, which also increases dust removal. This installation at a foundry enabled a steady blast temp. of 280–300° C. to be maintained and gave, with a reduced coke rate, iron at 1410° C. Further advantages claimed are: Productivity increased by 60%, and sulphur content of iron reduced.—S. K.

Experiences with Cupolas. J. Plch. (*Slévárenství*, 1954, **2**, (5), 144–146). [In Czech]. The principle of continuous outflow of cast iron and slag through their respective tapholes in the cupola is explained, and an account is given of the modifications in cupola design required with this process. The design is based upon consideration of the behaviour of two immiscible liquids in contact in a connected vessel, allowance being made for different pressures acting on the respective free surfaces, *i.e.*, air pressure in the cupola. The iron is collected in a fore-hearth, and the slag is granulated in a stream of water.

Installation of a Spray Cooled Cupola. F. Ferrat. (*Usine Nouvelle*, 1954, **10**, Mar. 18, 67). The construction of a water-cooled cupola is described, the refractory lining being only 3 cm. thick in the sprayed region. The advantages ensuing including flexibility of production, longer lining life, a better heat balance, and the possibility of using a higher percentage of steel scrap, are discussed.—A. G.

The Incidence of Metallurgical Factors on the Cost Price of Light Grey-Iron Castings. M. E. Doat. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–26). [In French]. The most suitable grey iron composition for thin castings is discussed and the following is suggested: C 3·1–3·5%, Si 2·5%, Mn 0·55%, S < 0·14%, P 1·10%. The choice of a cupola charge to give such a composition is considered from the points of view of reproducibility of the results, simplicity of operation, and minimum cost. Methods of economising in the use of ferro-alloys are outlined.—B. C. W.

The Hot-Blast Cupola and Its Applications. J. Estival. (*Génie Civil*, 1951, **131**, May 1, 170–172). The development of the hot-blast cupola as a producer of a combustible gas is traced. Such a cupola can yield a gas with a calorific value of 800–1100 kg.cal., or, if blown with oxygen and steam, it gives a gas suitable for ammonia synthesis. Other economic possibilities are in the treatment of scrap with low-grade coke or the use of previously agglomerated fine ore.—A. G.

The Influence of Casting Temperature on Chill and Mottle Formation. W. J. Williams. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5**, Feb., 136–144). The influence of casting temperature on the chill depth of metal melted at two different temperatures has been studied. With melting temperatures above a certain value, a decrease in casting temperature decreases the chilling tendency. Additions of aluminium accentuate the effect of casting temperature. In irons of very low sulphur content, variations in casting temperature give erratic chilling response. Rapid cooling in the ladle decreases the chilling tendency.—B. G. B.

Cupola Development (*Inst. Brit. Found.*, 1954, June 22–25, *Paper No.* 1089). This is the first report of Sub-committee T.S.43 of the Technical Council of the Institute of British Foundrymen. It reviews in detail British experiences with hot-blast, water cooling, carburizing, and desulphurizing in both basic and acid cupolas. Data are reported from a unit employing both hot-blast and water cooling, and particular attention is paid to the effect of hot-blast and basic slag practice on carbon pick-up. Among the conclusions reached are that the main advantages claimed for hot-blast practice are in fact achieved, and that water cooling offers a possible solution to the lining problem in basic cupolas and should be further investigated.—B. C. W.

Design and Operation of a Modified Cupola. F. A. Woolley. (*Inst. Brit. Found.*, 1954, June 22–25, *Paper No.* 1090). A modified cupola is described in which heat can be supplied to the contents of the well by two carbon electrodes projecting into the coke and supplying low voltage, high amperage, electric current. The power is used intermittently to smooth out irregularities that occur during melting. The process gives an improved supply of molten metal to the pouring stations. It also widens the range of raw materials that can be used in the furnace.—B. C. W.

Cupola Practice: Second Thoughts. W. Y. Buchanan. (*Inst. Brit. Found.*, 1954, June 22–25, *Paper No.* 1091; *Foundry Trade J.*, 1954, **97**, Sept. 16, 315–324). Twenty years' operating experiences with a 45-in. balanced blast cupola are reported and discussed. Details are given of methods of patching the cupola, which operation is done pneumatically with a moistened mixture of crushed firebrick and fireclay.—B. C. W.

Cupola Melting of Cast Iron Borings and Steel Turnings. W. Y. Buchanan. (*Amer. Foundryman*, 1954, **25**, May, 127–137). A detailed account is given of experiments carried out over the past 25 years in the cupola melting of cast iron and steel borings and turnings. The methods used have included bonding the borings with cement, continuous melting of the borings in a crucible to give small pigs, and mixing the steel turnings with molten cast iron in a rotating bowl to produce lumps suitable for charging into the cupola. For many years borings were successfully melted together with scrap or pig iron in a 32-in. balanced blast cupola, and recently loose borings or turnings have been melted in both 24 in. and 32 in. cupolas.—B. C. W.

Simple Foundry Expedients Proved Worthwhile in Recent Years. J. E. O. Little. (*Inst. Brit. Found.*, 1954, June 22–25, *Paper No.* 1098). Eighteen successful ideas which have been drawn from two different foundries are illustrated and discussed. The ideas cover all aspects of foundry operations from cupola practice to the cleaning of castings.

Experimental Manufacture of Spheroidal Graphite Chilled Iron Rolls in Hungary. B. Körös. (*Acta Technica*, 1954, **8**, (1–2), 37–66). [In Russian]. Trials in the manufacture of spheroidal graphite chilled iron rolls in Hungary by different methods are reported. Three types of addition were used: (1) A master alloy containing approx. Mg 22%, Cu 10%, Si 45%, balance mainly iron; (2) the alloy (1) together with electron scrap in such proportions that half of the total of 0·55–0·65% Mg added came from alloy (1) and half from the electron; (3) sufficient pure electron alloy to add 0·55–0·7% Mg. Addition (2) reduced the final silicon, increased the depth of chill and the hardness of the journal. Addition (3) proved to be the most efficient, and lighter rolls are now being produced regularly by this method.—R. A. R.

An Industrial Method for the Production of Spheroidal Graphite Iron. R. Radtke. (*Met. v. Giesserei Techn.*, 1954, **4**, May, 231–234). In the method described magnesium in rod form is continuously fed into the molten iron through an enclosed sleeve and a hole in the side near the bottom of the ladle. A satisfactory technique has been developed for producing spheroidal graphite cast iron.—R. A. R.

Formation of Graphite Nodules. A. De Sy. (*Metal Progress*, 1954, **66**, July, 92–93). The mechanism of the growth of graphite nodules in nodular iron is briefly considered. When the graphite particles are very small they start to grow within the solid austenite by a mechanism analogous to a peritectic transformation. This mechanism is capable of explaining several of the characteristics of nodular irons.

Modern Production of Whiteheart Malleable Iron. J. Bernstein. (*Inst. Brit. Found.*, 1954, June 22–25, *Paper No.* 1093). Details are given of the methods of producing whiteheart malleable iron castings in the foundry of the Simplex Electrical Co., Ltd. The castings are annealed in controlled atmosphere electric furnaces, and for heavy castings the annealing cycle is arranged so that the final microstructure consists of spheroidal carbides in a ferrite matrix, rather than the decarburized structure usually associated with whiteheart malleable iron.—B. C. W.

A Contribution to Improve the Cupola Malleable Iron. T. Shimomura. (*Tetsu to Hagane*, 1952, **38**, May, 296–301). [In Japanese]. A practice was investigated, using duplex melting with a 2-ton cupola and 4-ton reverberatory furnace. The mechanical properties of the product were markedly affected by the CaO/SiO_2 ratio of the cupola slag; when this was increased the graphitization cycle could be shortened and the properties of the annealed product were improved. Oxidation in the cupola should be minimized, and superheating in the reverberatory furnace is necessary.—K. E. J.

Malleable Cast Iron. (*Usine Nouvelle*, 1954, Spring Issue, 77–79). Two kinds of malleable cast iron can be obtained by varying the reheating conditions. In the first case, carbon is removed from the cementite by heating in an oxidizing atmosphere. Alternatively, the heat-treatment is carried

out at 700° C., in an inert atmosphere, causing precipitation of the carbon as nodular graphite. Mechanical and physical properties of the two irons are summarized.—T. E. D.

Malleable Iron Annealing Furnaces Converted to Natural Gas Firing for Improved Operation. (*Indust. Heating*, 1954, **21,** June, 1114–1116). Conversion of eleven 25-ton furnaces from coal to natural gas firing has enabled a reduction of 24 hr. in the firing cycle to be made.—R. W. B.

Malleable Iron—Recent Developments in Annealing Practice. P. F. Hancock. (Symposium on Heat Treatment, Liege, 1954: *Rev. Univ. Min.*, 1954, 9th series, **10,** July, 463–472: *Iron Coal Trades Rev.*, 1954, **169,** Aug. 20, 459–465). The practice of annealing of whiteheart and blackheart malleable iron is considered in detail. The type of gas atmospheres and furnaces used are described and the advantages of the new methods of annealing which have been developed in recent years are explained.—B. G. B.

Electric Furnaces in the New Laboratories of the British Iron and Steel Research Association. (*J. Four. Elect.*, 1954, **63,** Jan.–Feb., 19–21). A short illustrated description of the 500-kW. arc furnace and the H.F. induction furnace for vacuum melting, which have been installed at the Sheffield laboratories of B.I.S.R.A., is given.—B. G. B.

Steel Production on a Small Scale: Recent Developments in Oxygen Lancing Technique Lead to New Methods in the Foundry. D. J. O. Brandt and W. S. Williams. (*Iron Coal Trades Rev.*, 1954, **169,** Aug. 27, 516–518). A description is given of a method of small-scale production of steel developed jointly by Cross Foundry & Engineering Co. Ltd. and B.I.S.R.A., in which molten iron is refined in a ladle by oxygen lancing. The method is rapidly proving its value as a small-scale intermittent operation and has a great economic advantage in requiring a minimum of capital expenditure. It is at present confined to production of plain-carbon steel from phosphorus-free iron. The introduction of the process, and the refining and finishing operations, are described.—G. F.

Pre-Mixing of Reconditioning Materials for Moulding Sand. B. Jones. (*Amer. Foundryman*, 1954, **25,** May, 123–126). By mixing bentonite, cellulose, sea coal, and water before their addition to moulding sand a more uniform moulding mixture results with savings in the individual additions. Details are given of the conversion of a muller for mixing the binders, and operating experiences with the pre-mix are described.—B. C. W.

Theory, Economics, and Practical Application of Exothermic Materials. D. V. Atterton and C. Edmonds. (*Inst. Brit. Found.*, 1954, June 22–25, *Paper No.* 1092). The factors influencing the feeding of castings are briefly outlined and the methods available for delaying the freezing of feeder heads are considered. The only effective methods involve the use of either Thermit mixtures or the recently developed mouldable exothermic mixture. The relative technical and economic merits of these methods are discussed. The exothermic sleeves give a higher yield and a greater monetary saving, but Thermit method is more suitable for use on small thin sectioned castings where feeding is required at a large number of places.—B. C. W.

The Use of an Induction Furnace at Mains Frequency for the Production of Cast Iron. A. Tagliaferri and C. Barbazanges. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper* CI–50). [In French]. (*Fonderia Ital.*, 1954, **3,** Apr., 148–158). [In Italian]. The construction and operation of low-frequency induction furnaces are discussed together with the application in the foundry. Details are given of the cost of operation both with cold metal practice and with duplexing.

Keverian's Experiments on the Origin of Graphite Nodules. P. E. A. van Niewland. (*Metalen*, 1954, **9,** July 15, 203–207). [In Dutch]. The theory of Keverian and others on the formation of spheroidal graphite in high-purity Fe–C–Si alloys (*see J. Iron Steel Inst.*, 1954, **176,** Jan., 108) is confirmed by experiments reported in this paper.—R. A. R.

A Note on the Appearance of Spheroidal Graphite in Cast Iron. B. Marincek. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–25). [In French]. The solidification of both lamellar and spheroidal graphite cast irons is discussed with particular reference to the effect of surface tension. The formation of spheroidal graphite in magnesium-treated cast iron is attributed to the increase in the surface tension of the liquid iron produced by the deoxidation and desulphurization brought about by the magnesium. Increasing surface tension gives rise to greater undercooling and causes the graphite nuclei to grow in a form which has minimum surface energy.—B. C. W.

A Practical Study of a Casting System for a Pressure-Tight Casting in Spheroidal Graphite Iron. E. Mortara. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–29). [In French]. The problem of producing sound castings in spheroidal graphite iron is discussed from the points of view of casting temperature, casting system and correct feeding. The use of different types of feeder heads to reduce porosity is briefly considered and results are presented to show the effect of various feeder heads on the mechanical properties of cast test-pieces.—B. C. W.

Dilatometric Curves for the Determination of the Annealing Cycle for Spheroidal Graphite Iron. C. Pensotti and G. Pogatschnig. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–51). [In French]. Dilatometer curves from room temperature to 900° C. are given for spheroidal graphite, lamellar graphite, and malleable iron—both as-cast and after various preliminary treatments. The results are correlated with metallographic structure and it is concluded that with a ferritic-pearlitic structure a single treatment at 750° C. is sufficient to give complete ferritization.

Balanced Production of Malleable. A. R. Parkes. (*Found. Trade J.*, 1954, **96,** Apr. 15, 437–447). The plant and practice at Castings Ltd., Walsall, Staffs., is described with particular reference to the quality control techniques adopted for metal production, annealing and sand systems. The foundry averages 35 tons of finished castings a week, largely in blackheart malleable iron.—B. C. W.

Oxygen in Malleable Iron. L. W. L. Smith. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5,** Feb., 173–179). The influence of the oxygen content on annealability of white iron for the production of malleable iron has been studied, using the dilatometric technique. The influence of oxygen in the presence of aluminium has also been studied. In the absence of deoxidizing elements, oxygen in amounts from 15 to 110 p.p.m. does not appear to influence the time required to complete the graphitization of a white iron by annealing. Adding oxygen to a white iron containing aluminium increases the time required to complete graphitization.—R. A. R.

Pressure Pouring Steel Car Wheels. E. Q. Sylvester. (*Mech. Eng.*, 1954, **76,** Feb., 152–158). The author describes a new process for the casting of steel railway wagon wheels. They are cast in permanent graphite moulds and are pressure poured. The preparation of the graphite mould is fully discussed and the physical properties of typical wheels are given.—D. H.

Behaviour of Sand Mixtures Under Load at Room and Elevated Temperatures. R. G. Gooding and R. Rew. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5,** Apr., 278–295). An apparatus is described for obtaining stress/strain curves on dried compression test pieces of moulding sand at room and elevated temperatures. The machine is also capable of recording expansion curves on test pieces of the size used for physical tests. The effects of variation in moisture and clay contents on the stress/strain curves were obtained from the tests carried out at room temperature. The results of tests carried out at elevated temperatures on sand mixtures containing pitch and coal dust are compared with those obtained without these additions.—B. G. B.

Considerations on the Use of Sand Tests in the Foundry—A Suggested Stripping Test. F. Boussard. (*Fonderie*, 1953, Dec., 3707–3714: *International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* MO–8). [In French]. The properties required from moulding sand are briefly considered and it is pointed out that the results of laboratory tests frequently do not correlate with the behaviour of the sand in the foundry. A test is described which measures the force required to strip a cylinder of sand from the tube in which it has been compacted. The stripping force and strength are determined as a function of the moisture content of the sand for both synthetic and natural sands compacted to different degrees.

The Measurement of the Hardness of Moulds and Its Value for Assessing the Ramming of Sand in Moulding Machines. E. Piwowarsky and Patterson. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–41). [In French]. The hardness testing of sand moulds is discussed and the relation between mould hardness and the other physical properties of the sand is considered. A simple apparatus for testing mould hardness is described.

The Evolution of Gas at the Surface of Moulds and Its Determination. K. Roesch. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* MO–12). [In French]. Foundry and laboratory methods of measuring the quantity of gas evolved at the mould surface are described and results are reported of tests carried out with natural, synthetic, and oil core sands containing various amounts of moisture or sulphite lye. The significance of the results is discussed in relation to the correct choice of moulding sand, and the avoidance of casting defects.—B. C. W.

Recent Developments with Synthetic Resins in the Foundry Industry. P. G. Pentz. (*Found. Trade J.*, 1954, **96**, Mar. 25, 323–328). The use of synthetic resins in the shell-moulding process is discussed and a tentative theory is put forward to explain the phenomenon of 'peel-back.' The principles of the 'D' or Dietert process are outlined and the use of various types of binders in this process is discussed. The possible applications of cold-setting furane resins in the foundry industry are briefly mentioned.—B. C. W.

Gating Factors. W. H. Johnson, H. F. Bishop, and W. S. Pellini. (*Foundry*, 1954, **82**, Apr., 102–107, 271–272). The fundamentals of liquid flow in gating systems under conditions where there is no back pressure from the mould, are discussed and illustrated. The effects of pouring cup, sprue, and runner design on the flow velocity and volume rate of flow are considered, together with the problem of obtaining equal flow through all the ingates of a multiple gating system. Equal flow may be obtained by pressurizing, *i.e.*, choking at the ingates, or by enlarging the runner to develop a swirl in the metal flowing from the sprue, thus destroying its kinetic energy.—B. C. W.

Physical Experiments on Casting and Gating Practice. L. Frede and W. Magers. (*Giesserei*, 1954, **41**, May 27, 283–288). Using water and air models to show the fluid flow, numerous experiments are described to illustrate the physical laws governing casting practice. Flow speeds and flow processes in flat runners are described and illustrated in 36 diagrams. Streamline flow and turbulence are clearly shown in many cases. The different flow patterns obtained in a ring-shaped mould with tangential and with radial entries are compared. The different flow patterns incurred by casting from the top, bottom, or sides of moulds are also compared.—R. J. W.

Solidification and Dimensions of Feeder Heads. G. Berger and A. Belin. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–33). [In French]. A summary is given of Halbart's theory for the relation between the time of solidification and the volume/surface ratio of a casting, and experimental results are presented for bronze castings of simple shapes. A close correlation is shown to exist between T and V/S where T is the time of solidification and V and S are the volume and surface area respectively, of the casting. Halbert's theories for calculating feeder head size are considered and graphs are given for the minimum size of feeder head required for castings with given solidification times.

Contribution to the Study of Feeding Systems—Use of Exothermic Compounds. P. Nicolas. (*Fonderie*, 1953, Dec., 3715–3725: *International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–12). [In French]. The size, positioning, and effect of feeder heads are considered with special reference to the use of exothermic feeding compounds. Formulæ are derived for calculating the minimum weight of feeder head in relation to the weight of the casting and the minimum amount of exothermic compound necessary to ensure proper feeding.—B. C. W.

Two Methods of Moulding the Same Casting. H. Haynes. (*Found. Trade J.*, 1954, **96**, Apr. 8, 419–422). Alternative machine and hand ramming methods for producing moulds for cylinder liners are described and illustrated. By using a specially designed system the hand moulding was considerably quicker than machine moulding.—B. C. W.

Recent Developments in High-Pressure Moulding. T. Barlow and W. R. Adams. (*Foundry*, 1954, **82**, Apr., 111, 257–261). To obtain precision casting in green sand a uniform mould density is required and this can be achieved by regulating moulding pressure, sand flowability, and mould shape. The requirements in patterns and flasks for high pressure moulding are discussed, and the ramming action of flat platens and flexible diaphragms is explained. The flexible diaphragm exerts a uniform pressure on all parts of the mould and thus gives a uniform sand density.—B. C. W.

Centrifugal Casting on the Continent. (*Found. Trade J.*, 1954, **96**, Apr. 1, 359–361). A horizontal-axis centrifugal casting machine used for the production of cast iron pipes, and made by Ardeltwerke G.m.b.H., Wilhelmshaven, is illustrated and described.—B. C. W.

The Continuous Casting of Metals. Metallurgical and Mechanical Problems. (*Usine Nouvelle*, 1954, **10**, Mar. 25, 65–67). Techniques available for the continuous casting of metals are listed. The choice of method is discussed in relation to engineering, metallurgical and economic factors. Finally, processes suitable for various non-ferrous metals and for steel are described.—A. G.

Productivity in a Medium-Sized Foundry. J. Leonard. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–9). [In French]. The author does not consider that increased productivity can only be achieved by full-scale mechanization but suggests several ways in which a medium-sized foundry can be made more efficient by reorganization and careful planning.—B. C. W.

Improvements in Working Conditions in Foundries. Noise, Air-Conditioning, Lighting and Colours. J. Regnault. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–39). [In French].

Contribution to the Study of the Problem of Dust Removal from Cupola Gases. J. Prat. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–35). [In French]. A study was made with two normal cupolas of the concentration and particle size distribution of the dust in the outgoing gases. The particles smaller than 5 microns formed respectively 40% and 25% of the total, the lower figure being for the cupola with the lowest gas temperature. The exit gases from hot-blast cupolas were found to contain a much higher dust concentration and in the gases leaving the heat-exchangers the particles were almost all less than 5 microns in dia. Mechanical dust removal systems have an efficiency of 75–85% and for greater efficiencies electrostatic or wet precipitators must be used.—B. C. W.

Testing the Gas Content of Molten Metals. G. Ohira and V. Kondic. (*Foun. Trade J.*, 1954, **96**, Mar. 25, 331–333). A practical test for estimating the gas content of molten metals is described. The metal is slowly solidified in a mould designed on the Connor block principle and the density of the casting is found to bear a direct relationship to the gas content of the molten metal.—B. C. W.

Cold Cracks in White Iron. E. J. Jory. (*Foundry*, 1954, **82**, Apr., 122–123, 266–268). Cracks occurring at or near room temperature are the result of internal stresses set up on cooling. The mechanism by which the internal stresses develop is outlined and methods by which they may be minimized are considered. Casting and gating design is important but the most effective way of preventing cracking is by stress relief before the castings have cooled to room temperature.—B. C. W.

The Influence of Mould Factors on Casting Defects. J. H. Gittus. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5**, Apr., 264–267). The effects of casting dimensions and the composition of both metal and mould on the change in length of test castings during solidification are studied. Skin dried moulds and moulds containing unexpanded vermiculite tend to produce sounder castings than ordinary green sand moulds.—B. G. B.

Segregation of Manganese Sulphide Type Inclusions in Grey Cast Iron. R. Jolly. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5**, Feb., 180–186). The segregation of this type of inclusion is considered. An example of such inclusions in a cylinder casting is given and the results of previous investigations on manganese sulphide inclusions and experiments carried out by the author are discussed.—B. G. B.

A Cause of Cracking in Bath Castings. E. R. Evans and D. McK. Webster. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5**, Feb., 145–159). Cracking in bath castings in certain cases is due to the presence of abnormally high contents of antimony, boron, and lead in the iron. Field tests indicate that these elements originate from vitreous enamelled scrap included in cupola charges.—B. G. B.

HEATING FURNACES AND SOAKING PITS

Use of Plastic, Castable, and Gunning Materials in Soaking Pits. W. D. Rees. (*Iron Steel Eng.*, 1954, **31**, June, 59–63). The author describes the soaking pits at the Lowellville Works, Sharon Steel Corp. and discusses the advantages which have been derived by the use of plastic material for soaking-pit cover linings.—M. D. J. B.

New Soaking Pit Facilities at Steel Division of Ford Motor Company. B. D. Barnes and H. E. Raaflaub. (*Iron Steel Eng.*, 1954, **31**, May, 109–112). A brief description is given of the modernization and expansion programme at the Ford Motor Company's plant and details are given of their new soaking pits.—M. D. J. B.

The Heating of Steel. W. Trinks. (*Indust. Heating*, 1954, **21**, June, 1082–1086). This is an abridged English translation of a paper by W. Riemann on "Significance of Fuel Economy for Hot Working Operations in a Works Making Special Steels." (*See J. Iron Steel Inst.*, 1953, **175**, Nov., 336).—R. A. R.

The Use of Steel Recuperators in Rolling-Mill Reheating Furnaces. Trevoux. (*Centre Doc. Sid., Circ. Inform. Tech.*, 1954, (5), 955–974). Unlike refractory or cast-iron recuperators, those of steel can be made entirely air tight, permitting the use of pressure operation and facilitating control. Modern types of steel recuperator are described together with their automatic control and safety devices.—A. G.

Furnaces for Large Forgings. J. E. Lafon. (*Mét. Constr. Mécan.*, 1954, **86**, July–Aug., 585–591). The design and use of furnaces suitable for heating large ingots before forging are discussed. Fixed and mobile furnaces are considered together with suitable refractory bricks, methods of heating, and the organization involved in this type of work.—B. G. B.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Automatic Carbon Control. H. Ipsen and E. J. Rupert. (*Metal Progress*, 1954, **66**, July, 98–102). A description of the development of an automatic instrument for the control of the carbon potential of gases used in batch furnaces is given. The instrument is based on the measurement of dew point. Results are reported for carburizing and carbonitriding treatment using this method of control.—B. G. B.

Designing and Operating A Heat Treating Department. R. W. Wilson. (*Amer. Foundryman*, 1954, **25**, May, 102–106). The equipment and operation of a modern heat-treatment shop designed to carry out the annealing, hardening and tempering, stress relieving, and carburizing of castings, is described in detail.—B. C. W.

The Isothermal Heat-Treatment of High-Speed Steels and Its Effects in the Light of Recent Structure Research. H. Schrader and D. R. Dhanbhoora. (*Eastern Metals Rev.*, 1953, **6**, Apr. 13, 428–431). A review of methods of isothermal heat-treatment applicable to high-speed steels is made. This bainitic treatment is particularly suitable for low-alloy steels and the influence of alloy composition on the efficacy of the method is considered.—B. G. B.

Solution Heat Treating, Annealing and Strain Relieving in One Salt Bath Furnace. T. H. Atkinson and W. P. Thompson. (*Wire and Wire Prod.*, 1954, **29**, July, 744–745). An Ajax electric salt bath furnace is described, which serves for the solution heat-treatment of aluminium alloy wire, annealing of brass and copper wire, and the stress relieving of cold headed C1008 steel, $\frac{1}{4}$ in. in dia., cold-headed parts. The last treatment consists of 20 min. immersion in nitrate at 1050° F. followed by a hot water quench.—J. G. W.

The Isothermal Heat-Treatment of High-Speed Steels and Its Effects in the Light of Recent Structure Research. H. Schrader and D. R. Dhanbhoora. (*Trans. Indian Inst. Met.*, 1952, **6**, 54–95: *Eastern Met. Rev.*, 1953, **6**, Apr. 13, 428–431). The intentional formation of primary and secondary bainite and pro-eutectoid carbide during hardening and isothermal annealing before hardening are discussed. Primary bainite improves the continuous cutting efficiency of low-alloy high-speed steels while treatment to secondary bainite improves the toughness and interrupted cutting efficiency of steels having large proportions of residual austenite such as highly alloyed cobalt steels. The improvement in cutting efficiency is probably due to diffusion of alloying elements and enrichment of the residual austenite. A review of the possible alterations in carbide composition and their dependence on alloying elements and carbon content of the steel, annealing period in the pearlitic range, and holding time in the bainitic range, and the resulting influences, which control the behaviour of high-speed steels during heat-treatment has been given. An improving application of the pro-eutectoid carbide precipitation in the range 700–500° C. resulting in similar effects of alloying element concentration as the treatment to primary bainite has been suggested.—D. H.

Deboronization of Steels. J. C. Shyne and E. R. Morgan. (*Metal Progress*, 1954, **65**, June, 88–90). The results of an investigation into the deboronizing of steels during heat-treatment are discussed. Owing to loss of boron from the surface layers of steel during austenitizing, the hardenability of boron steels as measured by the Jominy test may be much too low unless the hardness measurements are made on flats ground deeper than the standard 0·015 in.—B. G. B.

A Comparison of Economics and Methods for Heating Steel Strip. C. E. Peck. (Amer. Inst. Elec. Eng.: *Blast Furn. Steel Plant*, 1954, **42**, July, 791–799). A detailed survey of the cost, advantages, and disadvantages of ten methods of continuously heating steel strip, is made. The methods considered are H.F. induction heating; 9600 cycle induction heating; heating by conduction of electric current; radiant heating using high temperature molybdenum, silicon carbide, Ni-chrome or metallic resistance elements; direct gas firing; and gas-fired radiant-tubes with or without a high temperature gradient in the furnace. A short description of each method is given.—B. G. B.

Heat-Treatment Furnace Developments. (*Metallurgia*, 1954, **49**, June, 283–300). A detailed account of recent progress in this field is given. A description and illustrations of a wide range of furnaces for annealing, reheating, stress relieving and normalizing are presented. Equipment for carburizing, tempering, and nitriding, as well as salt baths and induction furnaces, are also described.—B. G. B.

Hardening of Steel by Low Temperature Treatment. A. Niedzwiedski. (*Canad. Metals*, 1954, **17**, May 5, 49–51; May 20, 45–46). The practical advantages of keeping steel parts at − 120° F. for 4–6 hr. after conventional tempering are discussed and the benefits of using cooled cutting oils in the machining of metals is considered. (32 references).—B. G. B.

Protecting Atmospheres for Annealing of Thin Steel Sheet. J. Teindl. (*Hutnické Listy*, 1954, **9**, (7), 406–409). [In Czech]. After a theoretical introduction dealing with controlled atmospheres, a method for using dried, dust-free, desulphurized producer gas for bright-annealing hot rolled steel sheet is described. The equipment for cleaning the gas is considered and operating conditions are indicated.—P. F.

The Hardening of Cast Iron Roller Surfaces. L. Andrejew. (*Foundry News*, (Stalingrad), 1953, (1), 24–27: *Met. u. Giesserei Techn.*, 1954, **4**, July, 324–326). A new method for hardening cast-iron rollers is described. Rotary furnaces are used; a method has been developed suitable for hardening cast iron with the following composion: C 3·2–3·5%, Si 1·5–2·0%, Mn 0·7–0·8%, P 0·4–0·5%, S 0·12%, for rolls 280, 330, and 500 mm. in dia. The depth of the hardened zone was 15–30 mm.—L. J. L.

Some Thoughts and Remarks Concerning the Phenomena Accompanying the Tempering of Steels. A. Sourdillon. (Symposium on Heat Treatment, Liège, 1954: *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 414–430). A detailed review is made of the physical changes which occur during the heat-treatment of steels. The formation and decomposition of an austenitic structure, the effect of slow or rapid cooling, and variation in grain size are discussed in some detail. The results of recent research work are considered in relation to the varied types of transformation which occur in different types of alloy steels. Some of the experimental results cited are directly applicable to the improvement of tempering practice. (71 references).—B. G. B.

Synthesis of Heat-Treatment in a Gaseous Atmosphere. R. Boutigny. (Symposium on Heat Treatment, Liège, 1954: *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 430–438). The use of gas mixtures in heat-treating furnaces is discussed. The protection of steel against carburization or decarburization is considered, reference being made to the CO–CO_2 and H_2–H_2O equilibrium diagrams for the oxidation of iron. The use of special gas mixtures for cementation, carbon restoration, nitriding, carbonitriding, and gas pickling are discussed. Practical examples of the calculation of the composition of gas mixtures for specific purposes are given.

Heat-Treatment with Gases High in Nitrogen. J. Guingand. (Symposium on Heat Treatment, Liège, 1954: *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 438–445). The use of gas mixtures containing > 90% nitrogen for the heat-treatment

of steels is discussed. The production of such mixtures on an industrial scale from the combustion of a number of different gases is described. Data on the cost of these operations are given.—B. G. B.

Controlled Atmospheres, Based on Cracked Ammonia, in Heat-Treatment Furnaces, in Metallurgy, and in Mechanical and Electrical Construction. P. R. Duré. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 445–457). The preparation of gas mixtures of cracked ammonia and cracked ammonia with other gases are described. Their use for the treatment of a wide range of materials made from iron, alloy steels, and nonferrous metals and alloys is discussed. The optimum gas compositions for tempering, sintering, brazing, normalizing, galvanizing, and welding a wide range of materials are summarized in tabular form.—B. G. B.

Furnaces and Generator Equipment for Gaseous Carbon Restoration, Cementation, and Carbo-Nitriding. C. Demoulin. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 457–463). Continuous and batch furnaces for this type of treatment are described and line diagrams and photographs are shown. Equipment for the production of the special atmospheres required is considered briefly.–B. G. B.

TTT Curves and Thermal Treatment. L. Habraken. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 473–489). The isothermal transformation curves are first defined and then a detailed review of the type and conditions of formation of various structures in alloy steels is made. The practical use of the curves is explained and their limitations are considered. Over 30 photographs of steel structures are reproduced which show the different types of structures which can be found and the changes in structure which take place during heat-treatment. (27 references).—B. G. B.

Practical Application of Isothermal Heat Treatment. M. Stempfel. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 490–492). A short review of the application of isothermal transformation curves in heat-treatment is presented, reference being made to the tempering of a steel with eutectoid carbon and a self-hardening steel.—B. G. B.

Isothermal Tempering in Practice. R. Mossoux and E. Progneaux. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 492–503). Two modern examples of this method are considered in detail—the tempering of gas turbine blades and small arms parts. Data concerning the steel analysis, isothermal transformation curves and tempering practice are given. Methods for reducing the extent of deformation of the parts during the tempering cycle and obtaining the optimum properties from carbon steels without resort to costly alloying elements are discussed.—B. G. B.

Internal Stress Created by Surface Flame-Hardening. H. Bühler. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 504–507). The influence of several factors on the state of stress in specimens 45 and 58 mm. in dia. made from two types of steel (0·31 and 0·50% C) and flame-hardened under controlled conditions has been studied. The influence of the rate of advance of the tempered zone on the value and disposition of internal stress has been particularly studied and a relationship between the depth of tempering and internal stress has been obtained. (22 references).—B. G. B.

Surface Treatment by High Frequency. A. R. Baffrey. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 509–514). The mechanisms of the heating of a sample by external means and by high-frequency induction are considered. Example of the use of induction heating for the surface hardening of steels are given.—B. G. B.

Equipment for Gaseous Cementation. J. A. Swain. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 514–530). Gaseous methods of cementation and Carbonitriding steels are considered and equipment for carrying out these processes on both large and small scales is discussed in detail with the aid of numerous photographs. Gas generators, methods of control, and different designs of furnace are described.—B. G. B.

Modern Shaft Furnace for the Heat Treatment of Steels. H. Bühler. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 542–548).—A description is given of this type of furnace, particular reference being made to work done in Germany. Its advantages are pointed out.—B. G. B.

Conclusions. A. De Sy. (Symposium on Heat Treatment, Liège, 1954; *Rev. Univ. Min.*, 1954, 9th series, **10**, July, 548–552). The author briefly considers interesting points which have arisen during the conference. The mechanism of the structural changes occurring during tempering are discussed, reference being made to the time-temperature-transformation curves of two irons containing copper.

Protection of Hot Work Tools in Heat-Treatment. J. Y. Riedel. (*Metal Progress*, 1954, **65**, Mar., 85–87). The condition of many furnaces and molten baths is frequently carburizing and a method of heat-treating medium carbon (0·35–0·60%) hot work tool steels under these conditions has been developed. The results of a series of tests on the efficiency of a plating of copper over the tool steel to prevent carburization is discussed. Experiments were also carried out in furnaces having an ordinary air atmosphere—the test pieces after copper plating being packed in a variety of carburizing materials.—B. G. B.

Method for Improving Temperature Uniformity in Furnaces. O. Lutherer and R. J. Reed. (*Metal Progress*, 1954, **65**, Apr., 113–120). The results of an investigation into the distribution of temperature in heat-treatment furnaces are given. Suggestions for improving the distribution are made.—B. G. B.

Line of Salt Baths Heat-Treats 90-mm. Gun Recoil Mechanism. G. F. Hyde. (*Metal Progress*, 1954, **65**, May, 104–108). The heat-treatment of three large parts of this type of gun presented a number of difficulties owing to the awkward shapes which in some cases prevented machining after heat-treatment. Preliminary investigations in oil and salt baths are described. The salt-bath method was chosen and the practice which was finally developed is explained.—B. G. B.

Carbon Restoration of Bar Stock with a Batch-Type Furnace. H. W. Callahan. (*Metal Progress*, 1954, **65**, June, 95–99). The use of batch-type bell furnaces for the restoration of carbon in steel rounds, hexagons, flats, and squares from $\frac{1}{2}$ to 12 in. in dia. and coiled steel strip up to $\frac{5}{8}$ in. wide in coils up to 40 in. in dia. which has been decarburized during processing is described. Details of the furnace design and the gas generators are given.—B. G. B.

Flame Hardening of Heavy Forgings. S. Smith. (*Metal Progress*, 1954, **65**, May, 100–103). The flame-hardening of forged steel parts from 2 to 24 in. in dia. and up to 20 ft. long at the National Forge and Ordnance Co., U.S.A., is described. Details of the flame and quench equipment and the metallurgical state of the hardened surface are given.

On the Utilization of a New Gaseous Constituent C_2N_2O. M. Okamoto and N. Shirai. (*Tetsu to Hagane*, 1953, **39**, Dec., 1336–1342). [In Japanese]. A more simple and effective method of case-hardening and bright annealing using towns gas is proposed, by utilizing the cyanide salts and oxidizing constituents in a closed vessel. Steels were oxidized in towns gas, but bright-annealed or case-hardened by adding $K_4Fe(CN)_6$. The carbonitriding power was greatest with additions of 3 g./litre of gas, and this is attributed to the formation of C_2N_2O gas. (14 references).—K. E. J.

High-Frequency Surface Hardening of Rolling-Mill Rolls. J. Böhm. (*Hutnik*, (Prague), 1954, **4**, (5), 145–147). [In Czech]. The methods of application, advantages, and cost of high-frequency induction hardening of rolling-mill rolls in automatic equipment are discussed.—P. F.

Case Hardening of Steel by Carburising. M. B. Bever and C. F. Floe. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 63–84). The case hardening of steels by carburizing in solid, liquid, and gaseous media is described. The factors affecting case formation are discussed. Methods of achieving selective carburizing are given. The selection of steels for case-hardening, the heat-treatment of carburized steels, and the properties of the case are described.—J. E. J.

Case Hardening of Steel by Carbonitriding and Cyaniding. M. B. Bever and C. F. Floe. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 85–100). The general features of carbonitriding are discussed and some typical structures described. Variable factors in the process such as treatment cycles, atmospheres,

types of furnace, quenching, and tempering are discussed. An account is given of the hardness, hardenability, resistance to tempering, and wear properties of carbonitrided case.

Case Hardening of Steel by Nitriding. M. B. Bever and C. F. Floe. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 101–119). The nitriding process and furnaces are described, and the methods used for selective treatment are given. Nitralloy, stainless and high-speed steels, die steels, and nitri-cast iron are discussed, and some structures of nitrided cases are described and illustrated. The effect of process variables on case depth and hardness are discussed. The precautions taken to ensure dimensional stability in nitrided parts and the properties and applications of nitrided steels are stated.—J. E. J.

Dimensional Changes During Case Hardening. B. Finnern. (*Arch. Eisenhüttenwesen*, 1954, **25**, July-Aug., 345–350). Rings of carbon, chromium, manganese-chromium, and chromium-nickel steels have been carburized and hardened and the dimensional changes measured. These depend on the shape and type of material, but no clear relationship could be established with chemical composition, strength of core, surface hardness, and case depth. The effect of banding, if present, is very marked, dimensional changes being greatest in the direction of the banding. The general view that rings become smaller on hardening is not true; an increase in internal and external diameter is to be expected after hardening Mn–Cr steels in a hot bath. Banded materials tend to shrink. Simple hardening after carburizing produces the smallest deformation.—J. P.

Rapid Method for the Control of a Solid Case-Hardening Agent. Z. P. Gutovskaya. (*Zavodskaya Laboratoriya*, 1950, **16**, (11), 1398–1399). [In Russian]. A brief account is given of a rapid method for the determination of carbon dioxide and moisture in solid case-hardening agents.—S. K.

Case Hardening Steel by Carburizing. (*Steel*, 1954, **135**, July 5, 62–65). A general description of modern carburizing is given. The objects, methods, mechanism, suitability of steels, and subsequent heat-treatment are discussed.—D. L. C. P.

Nitriding of Austenitic Chromium–Nickel–Molybdenum Steel Tubes with High Pressure Nitrogen at 700° C. F. Braumann and H. Krächter. (*Arch. Eisenhüttenwesen*, 1954, **25**, July-Aug., 373–375). The nitrided layer on the inside of tubes, made from 0·1% C, 17·4% Cr, 2·6% Mo, 9·6% Ni, 0·6% Ti steel, after 10,000 hr. treatment at 700° C. with high pressure nitrogen has been examined by chemical, metallographic, and X-ray methods. The nitrogen content had risen from 0·017% in the untreated tube to 2·2% in the nitrided layer. This layer was brittle and showed fine hairline cracks. X-ray examination revealed that in the austenitic matrix there were present two nitrides, a mixed iron-chromium nitride (Fe,Cr)N and almost pure iron nitride Fe_2N.—J. P.

Selective Heat Treatment. S. L. Case and H. J. Grover. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 154–162). The principles of surface hardening by induction heating, ceramic burners, and flame-hardening machines are discussed. A comparison is made of the performances and economic considerations of the three processes.—J. E. J.

Modern Methods of Wire Annealing. (*Wire Ind.*, 1954, **21**, July, 704–707). The Birlec bright annealing plant for wire at Rylands Bros., Warrington, is described. It consists of three installations complete in 1937, 1943, and 1953 respectively. The first has two sets of four hearths and two furnaces rated at 180–185 kW. The second is for high carbon steel wire and consists of 32 hearths and 8 furnaces. The third is mainly for blue and bright annealing low-carbon steel wire from 20 to 3 gauge; for this purpose these are 12 hearths and 3 furnaces.—R. A. R.

Annealing and Patenting with Salt Baths. L. B. Rousseau. (*Wire and Wire Prod.*, 1954, **29**, Apr., 399–400). Recent improvements in salt bath furnaces are summarized and their advantages for the annealing and patenting of steel wire are indicated, with examples from American wire mill practice.

FORGING, STAMPING, DRAWING AND PRESSING

Cold Heading. F. C. Boyd. (*Machine Design*, 1954, **26**, May, 187–190). Improvements in tool design and the development of tungsten carbide dies have made much longer production runs possible on cold heading machines. Cold heading as a production process has many advantages. It is economic, there is virtually no scrap, the tensile strength of the metals is increased by cold working and surface finish is excellent for most purposes.—M. A. K.

Tapered Shells Require Skilful Operation Layout. S. R. Cope. (*Machinist*, 1954, **98**, July 3, 1178–1181). The article explains the application of certain rules which must be followed for successful drawing of tapered shells. Factors affecting the design of preliminary or stepped shells are discussed.—M. A. K.

New Extrusion Techniques Developed. H. Albers, G. Krause, and A. Greensite. (*Mech. Eng.*, 1954, **76**, Apr., 343–345). The production of longitudinally stepped and multi-shouldered extrusions and combination die forgings and extrusions are described with examples of their application to the manufacture of poppet valves, shells, and turbine blades. Automatic operation and improved presses are discussed.—D. H.

Straightening Extruded Shapes. L. Hoffmann. (*Mech. Eng.*, 1954, **76**, Feb., 139–142). Equipment is described for the straightening, stretching, and untwisting of extruded sections. The application of stretching to welded steel pipes whereby the pipe is expanded diametrically by internal fluid pressure is discussed.—D. H.

Extrusion of Steel: Manufacture of Rings for Jet Engines. Extrusion of Stainless Steel Sections. (*Engineer*, 1954, **197**, May 28, 797–798). These articles describe the manufacture of extruded flash butt welded precision rings in high-grade stainless heating-resisting steels for gas turbines. It is claimed that the material when manufactured in extruded lengths from rolled billets has much better properties than cast material. The method is the outcome of collaboration between the Chesterfield Tube Co. Ltd., and Reynolds Tube Co. Ltd.—M. D. J. B.

Curved Stretching-Jaws. (*Aircraft Prod.*, 1954, **16**, May, 192–199). The use of a machine for the stretch forming of sheet metal is described, and typical savings in materials and cost are given.—B. G. B.

The Economic Production of Cold Drawn Special Sections in Small and Medium Batches, with Reference to the Hot Rolled Initial Stock. K. Wenderlich. (*Stahl u. Eisen*, 1954, **74**, July 1, 876–880). A hot pre-rolling procedure is described which permits the economic supply of hot rolled stock in small and medium batches for the cold drawing of special sections. In addition, suitable drawing tools, the best way of making them, and the most convenient drawing process are indicated.—J. P.

General Principles for the Evaluation of Research Results in Plastic Deformation. H.-J. Kühne. (*Technik*, 1954, **9**, Mar., 143–148; May, 225–231). The author recommends that in evaluating the dependence of forces in extrusion, cold heading, and wire drawing, on process parameters, the yield stress of the metal and a ratio between characteristic dimensions of the stock should be used as common denominators, so that data may be generalized by means of empirical exponential formulæ.—J. G. W.

New Ideas for Wire Manufacture. N. Davidson. (*Wire Ind.*, 1954, **21**, July, 715–719). The author argues that technical difficulties and costs would be reduced by adopting a range of double holing machines running at moderate speeds and feeding into gravity blocks developed by Sir James Farmer Norton and Co., Manchester, in the place of modern multi-hole, high speed machines. He illustrates his argument by calculations of output.—J. G. W.

Opinion *vs.* Fact in the Wire Industry. A. U. Sternlof. (*Wire and Wire Prod.*, 1954, **29**, June, 639–640, 698). Quality control, as applied to the wire industry, is briefly described and advocated.—J. G. W.

Modern Five-Pass Back-Pull Wire Drawing Machine. (*Wire Prod.*, 1954, **3**, June, 8–9). A machine invented by Rehnqvist and manufactured by Kilshammars Mekaniska Verkstad is described. It is of the non-slip, non-accumulative type, the speeds of individual blocks being matched automatically by differential gear transmission. The machine is very compact, with three capstans on one axis, and the remaining two on another, and with one motor driving all five. Heavy back-pull is employed, and capstans and dies are water-cooled.—J. G. W.

Practice and Control for the Manufacture of High-Carbon Commodity Wire. R. H. Isenberg. (*Wire and Wire Prod.*, 1954, **29**, June, 625–627, 666–668). Manufacturing and

inspection practice at the Pittsburgh Steel Company are described for high-carbon steel wire for springs and for steel-reinforced aluminium conductors. Stress is laid on steel making, rolling, and patenting practice. The detrimental effect of the absence of water cooling in the wire drawing machines on the mechanical properties of the wire, is illustrated by an example.—J. G. W.

Stainless Steel Wire Properties Are Affected by Draw Speed. S. Storchheim. (*Wire and Wire Prod.*, 1954, **29**, May, 522–523, 564–565). Investigations of the effect of drawing speed on the physical properties of hard-drawn stainless steel wires are reported upon. Considerable increases of coercive force and decreases of remanence with increasing speed were recorded, whilst the tensile strength and specific resistance decreased a little. Stainless steel (18/8) with an incorrect analysis for a particular specification of final properties may be made to satisfy the specification by an appropriate change in drawing speed.—J. G. W.

New Method of Measuring the Bearing Length of Drawing Dies. W. Lueg. (*Stahl u. Eisen*, 1954, **74**, July 1, 874–876). Recent developments in apparatus for measuring the bearing length of drawing dies are reviewed and a new method is described. In this, one end of the bearing length is adjusted to lie on the axis of rotation of a supporting table, so that when the table is turned through 180°, the position of the other end of the bearing length is easily determined by means of an optical or mechanical member. Two instruments employing this principle are described.—J. P.

The B.I.S.R.A. Diamond Die Profiloscope. (*Wire Ind.*, 1954, **21**, May, 518). A new instrument for the inspection of the bore and measurement of tapers of diamond wire-drawing dies is described and illustrated. One model covers die bores down to 0·010 in., and the other down to 0·002 in.—J. G. W.

ROLLING-MILL PRACTICE

A New Theory of Hot Rolling. J. S. Mican. (*Iron Steel Eng.*, 1954, **31**, May, 55–71). The fact that knurled rolls placed in a slabbing mill reduced surface tears in the slabs, roll wear, breakages, and power consumption, prompted the author to seek a plausible explanation and develop a new theory of hot rolling. A new method for the construction of hot rolling pressure diagrams is derived in detail and numerous pressure diagrams are given. It is claimed that progressive increases in the coefficient of friction result in decreases in rolling pressures and decreasing trends for roll wear and surface tears. It is also claimed that surface tearing decreases with increase in roll diameter. (12 references).—M. D. J. B.

The Quality of Recent Cast Iron Rolls. K. Otani. (*Tetsu to Hagane*, 1953, **39**, Oct., 1189–1202). [In Japanese]. A survey is given of the developments in materials for rolls in Europe, America and Japan; it includes plain and alloy chilled rolls, compound rolls, sleeve chilled rolls, centrifugally cast chilled rolls, cast carbide chilled rolls, grooved grain and chilled rolls, and the heat-treatment of roll necks to prevent breaking. (61 references).—K. E. J.

New Slabbing Mill Speeds Steel Production Cycle. W. G. Patton. (*Iron Age*, 1954, **173**, Mar. 4, 151–155). A description is given of the Great Lakes Steel Corporation's new slabbing mill. This is a 45 × 90 in. universal mill which can reduce 20-ton ingots of 94 × 82 × 25 in. into slabs of 4½ × 76 × 182 in. in 2·4 min. Manipulators are fitted only on the exit side of the mill and only one turn is given to the ingot during the rolling operations. Completely automatic soaking-pit furnaces and a scarfing machine are included in the new equipment installed.—A. M. F.

Temperature Stresses in Iron Work Rolls. C. F. Peck, jun., J. M. Bonetti, and F. T. Mavis. (*Iron Steel Eng.*, 1954, **31**, June, 45–57). This paper describes a numerical method for calculating temperature stresses in iron work rolls during rolling. The analysis, based on the computed temperature distribution, shows stresses of 6000 lb./sq. in. tension radially, 70,000 lb./sq. in. compression tangentially, and 4,000 lb./sq. in. for the maximum shear on tangential and radial planes. These stresses are of sufficient magnitude to play an important part in the explanation of spalling.—M. D. J. B.

Modern Light-Section Rolling Mill. L. Wegmann. (*Demag News*, 1953, (134), 5–10). A description of the design of a new rolling mill is given; it has a comparatively small rate of production (2½ tons/hr.) but is specially designed for flexibility, economy in production, and low initial costs. Angles, **T** bars, rounds, squares, and flats can be rolled.—B. G. B.

Spreading in the Flat Rolling of Round Wire and the Determination of the Shape of the Cross Section. W. Lueg and K-H. Treptow. (*Stahl u. Eisen*, 1954, **74**, July 1, 881–888). In order to confirm earlier investigations of the spreading during flat rolling of wire, four steels containing 0·03–0·90% C and of different diameters, were rolled in light and heavy passes with lubricated rolls. Particularly at low carbon contents, spreading increased with carbon content and with pass reduction. These effects are so great that they must be taken into account if the advantages of producing thin strip by flat rolling of wire are to be exploited. The deviations from the empirical rule $d_0/(b + h) = 0{\cdot}5$ are greater with lubricated than with dry rolls. In this formula, d_0 is the diameter of the original wire, b is the width and h is the thickness of the strip after reduction. The edge radius was determined for various rolling conditions. There is an obvious relationship between the radius r, the difference a between the outside width of the strip and the width in contact with the rolls, and the decrease in thickness. An equation for determining a is developed and some approximate methods for calculating r and a are given.—J. P.

The New Wire and Billet Mills at N.V. Nederlandsche Kabelfabrieke, Delft. G. Leder. (*De Ingenieur*, 1953, **65**, (41), 155–166; (46), 185–200: *Stahl u. Eisen*, 1954, **74**, Aug. 12, 1076–1079). The layout design, electrical equipment, and operational procedures at the new plant at Delft are described briefly. Features of particular interest are illustrated.—J. P.

The Park Gate Iron and Steel Company Limited: The 11-in. Continuous Bar Mill. (*Iron Coal Trades Rev.*, 1954, **168**, June 25, 1523–1554: *Engineering*, 1954, **177**, June 25, 818–821: *Iron Coal Trades Rev.*, 1954, **168**, June 25, 1523–1552: *Brit. Steelmaker*, 1954, **20**, Aug., 310–318). A detailed description is given of the plant and ancillary equipment of the 11 in. 17 stand continuous bar mill of the Park Gate Iron and Steel Co., Ltd., incorporating the most modern features. An account is first given of the buildings and civil engineering work, including preparations of the site, followed by details of the two continuous reheating furnaces, two-zone-fired, and of their instrumentation. This is followed by a full description of the mill and its mechanical and electric equipment, with layout diagrams and colour photographs. Finally, the services of the new plant are described.—G. F.

Mills at Fairless Use New Methods. G. E. Farrington. (*Iron Steel Eng.*, 1954, **31**, June, 125–126). The author describes a new method of descaling strip which involves the use of a 2-high temper mill, driven by a single 350 h.p. motor installed directly in the pickle line. He also describes an innovation on the slitting and shearing line.—M. D. J. B.

Pittsburgh Steel Complete Cold Mill Facilities. (*Iron Steel Eng.*, 1954, **31**, May, 141–142). A description is given of the new cold rolling mill of the Pittsburgh Steel Co. at Allenport, Pa. The mill is a 4-stand, 4-high, 66 in. tandem mill designed to operate at delivery speeds up to 3170 ft./min. Extra heavy housings will permit the mill to be used for rolling stainless steel or tinplate.—M. D. J. B.

MACHINERY FOR IRON AND STEEL PLANT

A Power System Expands with Steel Making Capacity at Detroit Steel. L. DeFiore and M. N. Hickok. (*Blast Furn. Steel Plant*, 1954, **42**, July, 767–777). A detailed account is given of recent electrical equipment installed by the Detroit Steel Corp. at Portsmouth, Ohio. Blue prints which show the layout of the plant, overhead and underground electrical cables, and the special electrical facilities required for the operation of an integrated steelworks are reproduced.—B. G. B.

Motorized Tongs for Coil Handling. (*Brit. Steelmaker*, 1954, **20**, June, 212–213). Brief details are given of the specially designed motorized tongs, carried by an overhead crane, which have been installed at the Cleveland, Ohio, works of the Jones and Laughlin Steel Corp. to operate between the continuous pickling line and the cold reduction mill. Coils from 20 in. to 72 in. wide, and up to 60,000 lb. in weight, can be carried without damage, with greater safety, and with a saving in time.—G. F.

Intra-Plant Handling of Round Billets and Pipe with Special Automotive Equipment. J. D. Tyson. (*Iron Steel Eng.*, 1954, **31**, June, 107–114). This article describes material handling by fork and straddle trucks inside the Lorain plant of the National Tube Division of U.S. Steel Corp. The combination of straddle and fork trucks and very high frequency control makes for a high degree of efficiency in all handling operations.

Rail Transport Wagon Shunting Machines. J. G. Lee. (*Engineer*, 1954, **197**, Apr. 23, 612–614; Apr. 30, 651–653). Locomotives are used at nearly all British ports at which iron ore is imported. The advantages and disadvantages of other machines designed for shunting and hauling wagons are presented and the number required and their cost is discussed. The diesel shunting tractor appears to be the most versatile machine considered. Economically it is surpassed only by the capstan for shunting alone. For hauling duties the 150 h.p. diesel-mechanical locomotive appears to be the most suitable.—R. A. R.

Comparison of European and American Straightening Machines. M. Petitfrère. (*Centre. Doc. Sid. Circ. Inform. Tech.*, 1954, (1), 179–180). Modern straightening mills have 7 or 9 rollers whose spacing depends on the weight of section treated. A recent American installation is described permitting adjustment in all three dimensions. This has advantages but the degree of work hardening may be appreciable if the rollers are close together.—A. G.

Innovations in Forging Cranes. H. Ernst and K. Kleinschmidt. (*Stahl u. Eisen*, 1954, **74**, July 29, 1001–1006). Developments in the design of forging cranes and manipulators are reviewed, and the electrical equipment and the control of cranes from the forging press are described. It is shown that a satisfactory solution of the problem of remote control may be obtained by using alternating currents of sonic frequency. A preliminary and successful trial of the equipment on a pilot plant is reported.—J. P.

Crane Power Limit Switches. R. W. Graham. (*Iron Steel Eng.*, 1954, **31**, May, 98–102). The author describes work which was carried out in an attempt to standardize crane power limit switches. Recommendations are made relating to design improvements, operating techniques, and maintenance procedure with a view to reducing failures.—M. D. J. B.

A Safety Limit Stop. R. J. Harry. (*Iron Steel Eng.*, 1954, **31**, May, 104–108). A mechanism is described which provides a mechanical means to stop hook blocks without accidents. The limit stop consists of a contactor or circuit breaker, mechanically tripped, which is used to disconnect the motor, and a group of springs, located on the hook block or on the trolley which are used to absorb the kinetic energy of the revolving parts of the hoist, consisting of the armature, brake wheel, gearing drum, and shafting.—M. D. J. B.

Air Conditioning of Crane Cabs and Control Pulpits. H. Spitzer. (*Stahl u. Eisen*, 1954, **74**, July 15, 952–955). The introduction on an experimental basis of air conditioning of crane cabs and control pulpits was begun some time ago in German steelworks. The results were generally satisfactory and further installations have been erected. Endeavours should be made to reduce space requirements and to make the equipment more readily fitted to existing cubicles. In order to reduce the amount of cooling necessary, there must be good thermal insulation, effective radiation protection, and heat-absorbing glass. Cabins subjected to moderate or intermittent radiation can be protected by this glass.—J. P.

LUBRICANTS AND LUBRICATION

Lubrication. R. W. Clark, J. H. Fuller, and J. A. Gleber. (*Chem. Eng.*, 1954, **61**, July, 211–228). The mechanism of lubrication and the constitution of lubricants are discussed. Methods of lubricating plain and roller bearings, gears, and sliding surfaces are described. The necessity for lubrication scheduling is emphasized.—T. E. D.

The Function of Lubricants and Process Oils in Steel and Tinplate Works. W. Williams. (*J. Inst. Prod. Eng.*, 1953, **32**, Dec., 535–554). A comprehensive account is presented of the methods of lubrication used in steel and tinplate works in South Wales. Details of old and modern rolling mills, with particular reference to the type of bearings adopted, are given. Methods of lubrication using recirculation of oil and hydraulic systems are considered.—B. G. B.

Molybdenum Disulphide, a Novel Lubricant. G. Spengler. (*Z.V.d.I.*, 1954, **96**, June 11, 506–512; July, 683–687). The occurrence, extraction, structure, chemical and physical properties of molybdenum disulphide are outlined. Its frictional behaviour at low and high sliding speeds and the effect of oxidation on it are discussed. The role of molybdenum disulphide as an inhibitor of fretting corrosion is surveyed.—J. G. W.

Lubrication of Gears. (*Sci. Lubrication*, 1954, **6**, Jan., 13–17; Feb., 10–13; Mar., 25–31; Apr., 10–14; May, 22–26; July, 12–14). The basic elements of the mechanical operation of gears and of their lubrication are described. Operating temperature is an important factor. Properties required in gear lubricants include suitable viscosity, resistance to oxidation, corrosion resistance, high film strength, adherence to the surface, extreme pressure properties (with hypoid gears), and demulsibility. For open gearing, adhesion and resistance to moisture are the most important lubricant properties; the procedure for applying the lubricant is described. Descriptions are given of examples of large gear units lubricated by spray for cooling.—K. E. J.

WELDING AND FLAME-CUTTING

Heavy Armoured Vehicle Fabrication. J. W. Brind. (*Welding and Metal Fab.*, 1954, **22**, May, 166–173). The production by welded fabrication of the Centurion Mk. III armoured fighting vehicles is described. The turret is an armour steel casting with an armour roof plate welded into position after machining, and the gun mounting consists of a fabricated cradle welded into the armour steel mantlet casting.—U. E.

Atomic Hydrogen Welding in the Chain-Making Industry. A. T. P. Black. (*Welding and Metal Fab.*, 1954, **22**, May, 174–177). The atomic hydrogen welding process as applied to the fabrication of steel chains is described.—U. E.

Manufacture of Steel Chains and Their Maintenance. (*Usine Nouvelle*, 1954, **10**, May 20, 39–40). Methods of manufacture of chains are outlined, including hand forging and welding, and types of electric welding. Heat-treatment and testing of chains are discussed. The necessity for reheating chains to 900°–920° C. for 30–40 min. is emphasized. This ensures the reforming of the original grain structure which deteriorates as a result of stresses during service. The reheating is necessary at intervals of 4–10 months depending on severity of service.— T. E. D.

Resistance Butt Welding of Steel Sheets by the Foil-Butt-Seam Welding Process. F. Busse. (*Schweissen u. Schneiden*, 1954, **6**, June, 248–253). The principle of the foil-butt seam welding process is described. Thin uncoated foil is laid over the two edges of the sheet to ensure a uniform current flow through the two sheets which will result in a firm welded joint. The properties of the welds are reported.—U. E.

Weldability of a B–Mo Steel Related to Properties of the Heat-Affected Zone. C. L. M. Cottrell. (*Brit. Welding J.*, 1954, **1**, July, 315–321). The transformation characteristics and mechanical properties of a commercial low-carbon boron–molybdenum steel have been studied. Continuous-cooling transformation diagrams are given for steel with and without previous supersaturation with hydrogen.—U. E.

Army First Pass Groove Weld Grade Susceptibility Test. Z. J. Fabrykowski. S. Goodman, and B. A. Schevo. (*Welding J.*, 1954, **33**, Apr., 168s–172s). The Army First-Pass Groove Weld Cracking test is described. The test is related to groove welds as they exist in production welds, and based on the fact that first pass welds are more difficult to produce successfully. Restraint in the specimen is provided by preparing a slot in a single plate. Austenitic and ferritic electrodes are used. An application of the test is described.—U. E.

Field-Weld Test for Mild Steel Welds at Low Temperatures. J. R. Watt and J. J. Smetana. (*Welding J.*, 1954, **33**, Feb., 79s–82s). Three types of electrode were evaluated by laying a bead of weld metal transverse to the rolling direction on a 0·24%-carbon steel plate ¼-in. thick. Specimens ¼ × ¼ in. in section were cut from this and impact tested at + 94°, − 3°, − 50°, and − 109° F. with the weld metal in tension.

Maintaining the Corrosion Resistance of Welded Stainless Steel. J. P. Moore. (*Corrosion Technology*, 1954, **1**, June, 92–96). The author discusses the difficulties involved in welding stainless steel materials in such a way that corrosion resistance is preserved in the weld metal and adjacent areas. It is suggested that this would be achieved where parent metal and welding rods are very low in carbon.—L. E. W.

Ductile Cast Iron—Properties and Weldability. H. Edel. (*Schweisstechn.*, 1954, **8**, Apr., 45–46). The properties of ductile cast iron and its weldability are discussed. It can be welded without preheating with a special electrode Castolin No. 25, and it can be cut with electrode Castolin No. 01, without using oxygen, and planed with electrode Castolin No. 03.—R. A. R.

Hardfacing of Compound Cast Rails. A. Buchholtz. (*Schweissen u. Schneiden*, 1954, **6**, June, 238–240). The metallurgical properties of hard-faced compound-cast rails

are discussed. These rails have a head of low-alloy steel (0·7% C, 1·04% Cr), and the rest is of mild steel (0·2% C). The rails are hard-faced in two layers by electric welding using ferritic lime-coated electrodes. The intermediate layer contains 0·11% C, 0·58% Mn and 0·02% Mo, and is of high ductility, while the top layer containing 0·23% C, 1·05% Mn, and 0·92% Cr, will have similar properties to those of the original head.—U. E.

Hard Facing. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953 : Amer. Soc. Metals, 1954, pp. 163–173). The application of hard alloy deposits by welding is described. The methods of oxy-acetylene gas welding and electric arc welding are discussed in relation to hard facing.

Tungsten Carbide Hard Facing. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 180–190). The method of producing tungsten carbide in granules suitable for hard-facing and the means of applying the deposits are described. Granule hardness, hot hardness, the effect of welding methods on abrasion resistance, and the resistance of the composite welds to low stress abrasion are discussed.—J. E. J.

Hard Facing for Hot Wear Resistance. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 274–289). The general problems of hot wear and the requirements and thickness of resistant deposits are discussed. The equipment for and methods of application of such deposits by oxy-acetylene welding, manual electric arc welding and automatic submerged and open arc welding processes are described. Some particular hot-wear applications of air-hardening alloy steels, martensitic and austenitic alloy irons, high-speed steels, cobalt-base alloys and nickel-base alloys are given.—J. E. J.

Evaluation of Notch-Bend Specimens. P. P. Puzak and W. S. Pellini. (*Welding J.*, 1954, **33**, Apr., 187s–192s). The energy, fracture appearance, and ductility characteristics of Charpy **V**-notch, Charpy keyhole and Schnadt specimens were evaluated with respect to relative significance and practicability for correlation to service performance of welded structures.—U. E.

Factors which Influence Weld Hot Cracking. W. R. Apblett and W. S. Pellini. (*Welding J.*, 1954, **33**, Feb., 83s–90s). A hot cracking test was developed which evaluates the hot cracking susceptibility of weld deposits in terms of the severity of the mechanical conditions which are required to produce hot cracking.—U. E.

Effects of Peening Last Pass of Welds. W. S. Pellini and E. W. Eschbacher. (*Welding J.*, 1954, **33**, Feb., 71s–76s). Explosion bulge tests were conducted to determine the effects of peening the last pass of butt welds of ship plate. The performance of the welds is severely impaired by peening owing to embrittlement of the weld metal. It is concluded that peening of the last pass is a hazardous practice if the welds are expected to serve at temperatures low enough for the parent metal to be susceptible to brittle fracture.—U. E.

Peening Investigations. (*Welding J.*, 1954, **33**, Apr., 206s–208s). This is a report by the Peening Committee of the Welding Research Council. The recommendations on the peening of welds included: (1) Peening of the first and last passes of a weld is a dangerous practice; (2) Plates should not be peened at temperatures in the 500–900° F. range as this causes a pronounced increase in brittleness.—R. A. R.

MACHINING AND MACHINABILITY

Developments in the CeDeCut Technique. (*Machinery*, 1954, **84**, May 14, 1033–1040). An engineering workshop specially for CeDeCut investigations has been established at the Central Research Establishment of The Distillers Co., Ltd., Epsom, for exploring more fully the possibilities of liquid CO_2 as a cooling medium for turning and boring Nimonic and other heat-resisting materials.—M. A. K.

Lodge and Shipley Floturn Metal Forming Lathes. (*Machinery*, 1954, **84**, June 11, 1232–1236). Special lathes have been built by Lodge and Shipley Co., Cincinnati, Ohio, for carrying out forming operations by the flow turning technique. Thin-walled, hollow parts of parallel, conical or profiled shapes, and with varying wall thicknesses, are produced from thick flat blanks, shallow drawn cups or forgings. Considerable improvement of the physical properties of the metal results from the extensive cold working.—M. A. K.

Machining Hard Materials. J. Hinnüber and O. Rüdiger. (*Werkstatt u. Betrieb*, 1954, **87**, (2), 53–57: *Aircraft Prod.*, 1954, **16**, July, 260–265). A review of chemical, electrical, and ultrasonic vibratory techniques for machining, is made. (27 references).—B. G. B.

Peeling Shafts with Carbide Cutting Tools. A. Mackert. (*Stahl u. Eisen*, 1954, **74**, July 1, 869–873). Earlier investigations on the peeling of steel shafts are briefly reviewed. The present tests were carried out with carbide tools set tangentially to the shaft and were intended to indicate the maximum possible feed and tool life, and to determine the conditions for obtaining the best surface finish. The designs of the peeling machine and the tools are described and the results of bright and rough finishing tests are reported.—J. P.

" Twin-Fluid Two-Way " Cooling in Grinding. G. Pahlitzsch and J. Appun. (*Z.V.d.I.*, 1954, **96**, Apr. 15, 320–328). This method, due to the first author, consists of passing one fluid, with the main task of lubrication, radially outwards through the grinding wheel, and applying another fluid for cooling tangentially to the circumference of the wheel.—J. G. W.

Ultrasonic Drilling. (*Aircraft Prod.*, 1954, **16**, June, 240–241). An account of Mullard ultrasonic drilling machines for a vibratory process using a boron carbide slurry is given.

CLEANING AND PICKLING

Metal Cleaning and Finishing Handbook. (*Iron Age*, 1954, **174**, July 29, F1–F34). This feature article contains 25 charts and tables, in many cases with extensive notes, presenting technical data on numerous processes for cleaning, descaling, pickling, etching, plating, and polishing metals; specifications are also included.—R. A. R.

The Progress of Mechanical Descaling of Rods for Wire-Drawing—A Symposium. (*Wire and Wire Prod.*, 1954, **29**, May, 501–510, 571–576). The proceedings of a symposium held in Chicago in November 1953 here recorded consist of reports on current investigations on mill practice by seven members of the Wire Association and of subsequent questions from the floor together with answers by the introductory speakers. Various designs of descaler are described and the problem of lubrication is discussed at length.—J. G. W.

Acid Pickling—a Closed Cycle ? W. Fackert. (*Stahl u. Eisen*, 1954, **74**, July 1, 888–894: *Iron Steel Inst.*, 1954, TMS 221). Modern high-speed strip pickling lines require high acid contents controlled within narrow limits. This is economically feasible only if the pickling line is run as a closed cycle from which ferrous sulphate and water are removed from the spent liquor and the necessary amount of sulphuric acid added before the liquor is re-introduced to the line. The amounts of ferrous sulphate, water, and acid to make this cyclic process possible can be calculated by the method described by Pearson and Bullough (*see J. Iron Steel Inst.*, 1951, **167**, Apr., 439–445). A spray unit has been developed for evaporating spent liquor. The amounts of water to be removed and acid added in order to restore a spent liquor to its original composition were exactly as calculated. A process has been worked out for converting ferrous sulphate monohydrate to sulphuric acid. It involves heating the monohydrate with steam and nitric acid vapours, trapping the distillate of sulphuric, nitric, and nitrous acids, distilling off the nitrous compounds and reconverting these to nitric acid by aerial oxidation.—J. P.

Bibliography on Electropolishing. (*Product Finishing*, 1954, **7**, Apr., 59–62). This bibliography, compiled by the Production Engineering Research Association of Great Britain, contains about 50 papers published since 1947.

PROTECTIVE COATINGS

Internal Coating of Pipe in Place. M. B. Grove. (*Corrosion*, 1954, **10**, May, 142–146). Methods of cleaning and painting old and new steel pipes internally using plugs moved by compressed air are described. Plastic lining of pipe by this method is discussed and operating procedures are given in detail.—J. F. S.

The Disposal of Toxic Industrial Waste. F. J. Knight. (*Metallurgia*, 1954, **49**, June, 279–280). A short survey is presented of methods used for the disposal of cyanide wastes

in America. Conversion to volatile hydrocyanic acid with sulphuric acid and oxidation to cyanate by chlorine, sodium chloride, or calcium hypochlorite are reviewed.—B. G. B.

Chromium Plating of Cutting Tools. M. I. Andrianov. (*Stanki i Instrument*, 1954, (2), 21–23). [In Russian]. The technology, application, and advantages of the chromium plating of tools are described. An electrolyte containing 150 g. CrO_3 and 1·5 g. sulphuric acid per litre at 55–58° C. is recommended with a c.d. of 50–60 amp./sq. dm. The thickness of the initial deposit is 0·003–0·005 mm. the maximum permitted in subsequent renovation being 0·02–0·04 mm. Methods of stopping off portions of tools during plating are described for several types of tool. Special attention is given to the plating of broaches. Chromium-plated tools in operation should be kept below 450–500° C. by effective cooling.—S. K.

Nickel Plating from the Sulphamate Solution. M. B. Diggin. (Fourth International Conf. on Electrodeposition and Metal Finishing, Apr. 23, 1954, Advance Copy No. 16: *Trans. Inst. Met. Finishing*, 1954, **31**.

Observations on the Effects of Pre-Plating Hot Galvanized Steel Wire. H. H. Passolt and A. T. Baldwin. (*Wire and Wire Prod.*, 1954, **29**, June, 632–634, 668–669). In order to reduce the thickness of the zinc–iron alloy layer in hot-dip galvanized high-carbon steel wire, the authors investigated the use of a preheater with flux-coated and with flux-coated and pre-plated (with zinc, tin, and nickel) wire, coupled with greatly reduced immersion times in the galvanizing bath. The method constituted an improvement on conventional treatment, in terms of coating adhesion and ductility.—J. G. W.

Zinc Plate on Cartridge Cases Meets Rigid Specifications. C. E. Fisher and D. F. Zlatnik. (*Iron Age*, 1954, **173**, Mar. 11, 135–139). Substitution of steel for brass made it necessary to develop a corrosion-resistant coating to protect the steel base material. A zinc plating process involving 23 operations is described. Concentration, freedom from contamination and operating temperatures of solutions are very important in maintaining high production and quality. The complete cycle is fully automatic.—A. M. F.

A Consideration on " Greyness " of Tin Plates. K. Yoshizaki, T. Ando, K. Ariga, and Y. Miyamoto. (*Tetsu to Hagane*, 1939, **39**, Oct., 1166–1170). [In Japanese]. Investigations on tinplate made from semi-killed and rimming steel concerned oxidation and the effects of chemical composition. The defect of greyness is attributed primarily to a thin oxide layer 1–40 μ thick. (18 references).—K. E. J.

Preventing Corrosion with Saran. (*Chem. Eng.*, 1954, **61**, July, 264–272). The use of the thermo-plastic ' Saran,' a vinyl-chloride vinylidene copolymer, for pipes and fittings in standard sizes as well as for coatings is mentioned. Its behaviour at different temperatures in the presence of 122 common chemicals at various concentrations is shown diagrammatically and its physical characteristics are listed.

A Laboratory Technique for the Electrodeposition of Manganese on Other Metals. Winifred A. Bell. (Fourth International Conf. on Electrodeposition and Metal Finishing, Apr. 23, 1954, Advance Copy No. 27: *Trans. Inst. Met. Finishing*, 1954, **31**.

Sprayed Metals. D. A. Watson. (*Machine Design*, 1954, **26**, May, 166–170). Sprayed metal is a metallurgical material having entirely different physical and chemical properties from those of original wire. It contains oxide, is harder, more brittle and more porous than its equivalent in cast or drawn metal. Major applications include salvage of worn, damaged or wrongly machined journals on rolls, and the spraying of components liable to corrode.—M. A. K.

Metallizing May Be for You. H. Vanderpool. (*Chem. Eng.*, 1954, **61**, July, 189–194). Protective coatings of sprayed metal supplemented by organic coatings are discussed. A so-called ' metallizing system ' consists of a 0·010 in. layer of aluminium on mild steel, followed by a silicone resin base, aluminium pigmented paint. This would protect the steel from SO_2 and SO_3 mixtures at 450–675° C. in contact sulphuric acid plants. Other systems are mentioned.—T. E. D.

Metal Spraying—Modern Developments and Properties of Deposits. N. Collari and L. Foschini. (*Calore*, 1954, **25**, July, 298–306). [In Italian]. A review is given of modern metallizing techniques. The different characteristics and properties of metallized surfaces are examined, in particular hardness and tensile strength. Microphotographs of sprayed metal coatings are given together with tables of technical data.

Electroplated Coatings. J. M. Hosdowich. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 41–55). The principles of electroplating are briefly discussed. A detailed description is given of chromium, nickel, and rhodium plating. Methods of application and the wear-resistance properties of the deposits are described.—J. E. J.

Diffusion Treatments for Wear Protection. M. B. Bever and C. F. Floe. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 120–130). The diffusion treatment of steel with chromium and silicon is described. An account of the reactions and mechanism of the chromizing treatment is followed by a description of the processes in current use, the structures obtained in the cases, treatment procedures, case properties, and the applications of the process. The siliconizing process is briefly described.

Metallizing. H. Vanderpool and A. P. Shepard. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 131–153). The types of gun used in metal spraying are described and the mechanics of the process are discussed. The mechanical properties of the sprayed coating, particularly hardness, are dependent on porosity, cohesion of particles, the amount and nature of any oxides present, and particle hardness. These factors are considered.—J. E. J.

Hard Facing Alloys. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 174–179). The problems of classification of hard facing alloys are discussed from the user point of view. Guidance is given in the use of the A.W.S.-A.S.T.M. subcommittee's classification by chemical composition.

Chromium Carbide Hard Facing. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 191–201). Two types of chromium carbide hard facing alloys are described: Austenitic types in which the austenite is stabilized partly by chromium and partly by manganese or nickel; and hardenable types, which are simple high-chromium irons capable of being hardened or softened by heat-treatment.—J. E. J.

Cobalt-Base Alloys. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 215–224). The ranges of cobalt-base hard-facing alloys are described and the effects of modifications of composition and welding practice on their hardness and abrasion resistance are discussed.—J. E. J.

Nickel-Base Alloys. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals 1954, pp. 225–230). The uses of five main types of nickel-base hard-facing alloys and the effects of variation of composition on properties are discussed. The five types are: nickel-chromium alloys, chiefly 80% Ni, 20% Cr; Ni–Cr–Fe alloys; Ni–Cr–Fe–Si–B alloys; Ni–Cr–Mo–W alloys; and Ni–Fe–Mo alloys.—J. E. J.

Economic Justification of Protective Coatings. C. C. Harvey. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 295–304). The author discusses the economic aspects of painting as an anti-corrosion protection, taking into account the corrosion rate of the metal, the metal thickness loss which is tolerable, the value of the metal lost by corrosion, the useful life of the article or structure as controlled by factors other than corrosion, and the cost of initial painting and the maintenance of the coating.

Surface Preparation and Pretreatment of Metals Prior to Painting. A. J. Liebman. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 305–345). The natural surface conditions of metals and the effects of atmosphere on these surfaces are discussed. Surface contaminants resulting from working, storing, and shipping of metals are listed. An account is given of the equipment used and the types of surface finish resulting from six methods of surface preparation. The methods are: Mechanical cleaning; flame conditioning; blast cleaning with abrasives, open or closed wet system equipment, suction feed blast equipment,

or wheel blasting equipment; chemical cleaning; vinyl metal pretreatments, and oil pretreatments.—J. E. J.

Organic Coating for Normal Services. A. J. Eickhoff. (*'Surface Protection Against Wear and Corrosion,' Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 346–365). Four theories of corrosion are outlined: The peroxide theory, the oxygen attack theory, the alkaline theory, and the electrochemical theory. The types of test undertaken at Sayville Exposure Station, New York, to estimate the expected life of paint films are described, and an account is given of the corrosion problems encountered when metals are immersed in fresh or salt water. The economics of painting for plant maintenance are discussed in general terms.—J. E. J.

Corrosion Prevention by Paint Maintenance in Severe Industrial Atmospheres. K. Tator. (*'Surface Protection Against Wear and Corrosion,' Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 366–377). Cost comparisons are made between painting and 'over-designing' for anti-corrosion purposes, and between the application of durable paint materials and the more frequent application of cheaper materials. The author concludes that while painting costs slightly more than 'over-designing' the possibility of premature disuse of a structure through factors other than corrosion favours the use of paint, and that the difference in the ultimate costs of durable and inexpensive paint is negligible.—J. E. J.

Specifications for the Painting of Metals. J. Bigos. (*'Surface Protection Against Wear and Corrosion,' Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 378–395). A review is made of current specifications covering surface preparation of metals, pretreatment of metals, procurement and application of paint, and the handling, storage and protection of the metals to be painted. The difficulties inherent in the lucid writing of specifications are discussed. Guidance is given on the use of current specifications.—J. E. J.

Cathodic Protection and Galvanizing. H. A. Robinson. (*'Surface Protection Against Wear and Corrosion,' Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 396–426). Two types of cathodic protection cell are described and the mechanism of cathodic protection is outlined. The sequence of operations essential for the successful practice of cathodic protection is described. Some applications of the method are listed and discussed. Methods of galvanizing are described together with a discussion on the mechanism of protection. An account is given of the performance of galvanized iron products in service.

Metallizing. H. S. Ingham. (*'Surface Protection Against Wear and Corrosion,' Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 427–442). Two types of metallized coating are described, one in which the metal alone is applied and the other in which an organic seal is applied on top of the metal coat. The particular applications of these types of coatings are discussed with examples.—J. E. J.

Electrodeposited Coatings. F. Ogburn. (*'Surface Protection Against Wear and Corrosion,' Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 443–452). The procedure for the electrodeposition of anti-corrosion coatings is briefly described. The physical, mechanical, and chemical properties of such coatings are discussed and examples are given of the uses of zinc, cadmium, chromium, nickel, copper, tin, precious metal, lead, and anodized coatings.—J. E. J.

One-coat White Enamel. J. Semple. (*Found. Trade J.*, 1954, **96**, May 6, 533–537). The composition, properties, and processing of the titanium and lithium-bearing titanium super-opaque single-coat enamels used on sheet iron are discussed.

Symposium on Porcelain Enamels and Ceramic Coatings as Engineering Materials. (*Amer. Soc. Test. Mat. Spec. Tech. Pub. No.* 153, 1954). This symposium, sponsored by A.S.T.M. Committee C–22 on Porcelain Enamel, was held at Atlantic City on July 2, 1953. The papers presented were:

Some Examples of the Functional Use of Porcelain Enamel and Ceramic Coatings for Steel. G. H. Spencer-Strong. (3–9).

Resistance of Porcelain Enamels to Weathering. D. G. Moore. (10–18).

The Chemical Resistance of Glass Fused to Steel. E. A. Sanford and O. J. Britton. (19–25).

Acid Resisting Properties of Porcelain Enamels. H. C. Wilson. (26–28).

Requirements for and Expected Benefits from the Application of Coatings to High-Temperature Components of a Jet Engine. A. C. Francisco and G. M. Ault. (29–41).

High-Temperature Ceramic Coatings As Applied to Aircraft Power Plants. B. L. Paris. (42–50).

The Industrial Processing of High-Temperature Ceramic Coatings. J. H. Terry. (51–60).

The Abrasion Resistance of Various Types of Porcelain Enamel. A. V. Sharon. (61–69).

The Resistance of Porcelain Enamels to Surface Abrasion As Determined by the PEI Test. J. T. Roberts. (70–77).

Torsion Test As an Aid to the Porcelain Enamel Industry. E. L. Hoover. (78–86).

The Strengthening Effect of Porcelain Enamel on Sheet Iron as Indicated by Bending Tests. E. E. Bryant. (87–92).

Tension Tests of Porcelain Enamelled Steel. W. A. Deringer. (93–100).

Effect of Temperature on the Electrical Resistivity of Several Ceramic and Silicone-Type Coatings. S. W. Strauss, L. E. Richards, and D. G. Moore. (101–108).

A Laboratory Evaluation of Ceramic Coatings for High Temperature Applications. Sara J. Ketcham. (109–118).

Guideposts in Selecting Porcelain Enamels and Ceramic Coatings—A Summary. W. N. Harrison. (119–122).

Micaceous Iron Oxide in the Composition of Paints for Intermediate and Finishing Coats for Metals and Iron Alloys. H. Rabaté. (*Peintures, Pigments, Vernis*, 1954, **30**, July, 572–573). The characteristics of micaceous iron oxide as a pigment are summarized. It is used in paints for intermediate coatings and for the finishing of multi-layer anti-rust coatings. The most recent methods for improving the physical stability of oil paints containing micaceous iron oxide are given.

POWDER METALLURGY

Powder Metallurgy and Alloying. A. E. Williams. (*Min. J., Annual Rev.*, 1954, 111–113). Developments in the use of powder metallurgy during 1953 for the production of sintered permanent magnet alloys, mixtures containing titanium carbide for service at high temperatures, and electrodes for alkaline accumulators are described.—B. G. B.

Bonding in Cermets. L. S. Williams and P. Murray. (*Metallurgia*, 1954, **49**, May, 210–217). The criterion for selecting promising cermet materials is the likely bonding behaviour between the metal and the ceramic phase. In the case of surface interaction, the surface tension of the liquid metal is of primary importance; various relationships between the surface tension of metals and other properties have been studied in order to be able to predict values for the refractory metals. For bulk interaction, the main factor involves bonding *via* solid solutions, and this aspect is also considered.—B. G. B.

A New Prealloyed Stainless Steel Powder. A. Adler. (*Prec. Met. Mold.*, 1954, **12**, May, 54–56, 80–81). The production of stainless steel powder, its physical properties, and subsequent powder metallurgy processes are described. Pressed stainless steel scrap is furnace treated to cause brittle carbides to precipitate at the grain boundaries thus facilitating breaking down into powder.—R. A. R.

Production and Properties of High-Purity Nickel Powder. F. A. Forward. (*Bull. Inst. Metals*, 1954, **2**, May, 113–116). A short description is given of a new process that has been developed to produce high-purity nickel powder from a Canadian ore. Powder can be produced to meet a wide variety of particle size specifications.—B. G. B.

Method Described for Evaluation of Lubricants in Powder Metallurgy. I. Sheinhartz, H. M. McCullough, and J. L. Zambrow. (*J. Met.*, 1954, **6**, May, 515–518). The authors describe a simple method by means of which the effectivity of lubricants used in powder metallurgy can be evaluated, and present a series of mathematical relationships which can be used to predict results with a minimum of experiments. Several materials are shown to be satisfactory die-wall lubricants for cold-pressing.—G. F.

PROPERTIES AND TESTS

A Note on the Preparation of Miniature Tensile Test Specimens. R. J. M. Payne. (*Metallurgia*, 1953, **48**, Dec., 315). A method of preparing small flat specimens with a gauge

length of ½ in. is described, full details of the mounting and machining fixture being given.—R. A. R.

On the Ring Crushing Test of the Ball Bearing Steel. K. Deguchi. (*Tetsu to Hagane*, 1953, **39**, Oct., 1150–1159). [In Japanese]. A theoretical relationship was established between the results of the bending fracture test of flat test pieces and the ring crushing test, and was modified to suit practical requirements. It is proposed to use the fracture test in future, stipulating that the bending fracture strength must exceed 6000 kg.—K. E. J.

Effect of Testing Temperature on Notched Impact Toughness of Constructional Alloy Steels Tempered at Low Temperatures. T. Kawai and M. Nishada. (*Tetsu to Hagane*, 1954, **40**, Jan., 38–42). [In Japanese]. Studies were made on heats of Ni–Cr steel, oil-hardened and tempered at various temperatures. Tempering within the range of embrittlement elevated the transition temperature, while tempering at 200° C. produced the lowest transition temperature. Suggestions for reducing low-temperature-tempering embrittlement concern the steel-making process, in particular deoxidation, dephosphorization, nitrogen fixation, and grain-size control. (18 references).

Variation in the Impact Strength of Carbon and Alloy Steel; with Reduction of Temperature. L. Scuri. (*Calore*, 1954, **25**, May, 232–238). [In Italian]. The author studies the impact strength in carbon and alloy steels at temperatures ranging from + 20° C. down to − 60° C. The effects which methods of casting, rolling temperatures, internal structure, grain size, heat-treatment, and amount of alloying elements have on the mechanical properties of these steels at low temperatures are examined in detail with a view to recommending the best types.—M. D. J. B.

Measurements of Internal Energy in Cold-Worked Metals. F. Eugène. (*Microtecnic*, 1954, **8**, (3), 127–134). Experiments are described to investigate the internal energy developed in cold working. The energy supplied was recovered as heat to within 1%, which could be accounted for almost entirely by radiation and conductivity heat losses. The residual internal or latent energy under the experimental conditions described is less than 0·01 cal. per gramme of metal, which is too small to take into account in the experimental study of metal cutting.—L. D. H.

The Deltameter, a New Pneumatic Instrument for Dimensional Measurements. (*Microtecnic*, 1954, **8**, (3), 158–160). The principle and mode operation of the Deltameter are described. This is a pneumatic instrument based on a constant inlet pressure and an outlet pressure depending on the dimensions of the article to be measured. The Delta valve, giving a large range and high sensitivity, is described, and examples of various modifications of the instrument and their application to continuous quality control are given.—L. D. H.

An Investigation of the Brittle Strength of Mild Steel by Means of Notched Tensile Bars. J. Short. (*Iron Coal Trades Rev.*, 1954, **169**, Aug. 13, 389–396). The author has carried out an investigation of the brittle strength of mild steels by means of notched-bar tensile tests to provide data for a theory of brittle fracture in low-alloy steels. The results are presented and possible dislocation theories are discussed. The method has limitations but is shown to yield useful results when proper regard is given to technique; for instance, it is deduced that technical cleavage strength has little dependence on degree of strain at a particular temperature. In general, a quenching treatment is found to confer greatest resistance to brittle failure.—G. F.

Behaviour of Rolled Steel Joists in the Plastic Range. J. W. Roderick and H. H. L. Pratley. (*Brit. Welding J.*, 1954, **1**, June, 261–275). A description is given of tests performed on 8 × 4 in. and 10 × 4½ in. steel joists tested as simply supported beams bending about the major axis, with either a central concentrated load or two equal loads equidistant from the centre. Observations were made of the deflections, strain, and occurrence of yield bands at loads up to those at which the deflections became considerable. Bending and tension tests were correlated by comparing experimental load/deflection curves for beams with theoretical curves obtained by the application of the plastic theory.—U. E.

Shear and Slip. R. Vallette. (*Génie Civil*, 1954, **131**, Aug. 1, 284–289). The theory of strain and deformation is outlined, and several examples of deformation and rupture by shear and slip are examined. The properties of concrete and steel are discussed in relation to these theories.—T. E. D.

Report on Brittle Fracture Studies. F. J. Feely, D. Hrtko, S. R. Kleppe, and M. S. Northup. (*Welding J.*, 1954, **33**, Feb., 99s–109s). Causes of storage tank failures are discussed and results are given of a tensile test for notched specimens. Under constant stress and temperature conditions, a hardened steel wedge is driven into the notch using an impact gun. The data obtained indicate that below a certain temperature the brittle breaking stress is constant, while above this temperature the brittle breaking stress increases markedly.—U. E.

Fractures of Crane Hooks and Ladle Hangers and Measures for their Prevention. G. Will. (*Stahl u. Eisen*, 1954, **74**, Aug. 12, 1062–1069). Some typical fractures of crane hooks and ladle hangers, observed in recent times, are described. The fractures were due to ageing-susceptible materials, reduced cross-sections, or to incorrect welding and heat-treatment. The results of temperature measurements on ladle hangers, with and without radiation protection, have proved the high efficiency of a radiation shield, without which the inner faces of the hangers can reach a critical temperature of 150–250° C. The results of American stress and temperature measurements on heavy hangers are reported. In some circumstances temperatures of up to 550° C. are to be expected on the unprotected surfaces of the hook shanks. On the basis of the present experiments, rules are laid down for choosing materials, design, manufacture, treatment, and supervision of hooks and hangers.—J. P.

Cracking of Low-Alloy Steels Caused by Abrupt Temperature Changes. W. Rädeker. (*Stahl u. Eisen*, 1954, **74**, July 15, 929–940). Cracks sometimes occur in boiler tubes and plates and other components subjected to temperature variations. Tests have been conducted by submitting steel specimens to a treatment involving repeated quenching from various temperatures, and reheating. It has been found that the production of cracks requires a certain amount of mechanical stress, caused by the volume difference between the heated and cooled sides of the material. Cracks are not produced on quenching from 300° C. in water but they are quite marked on quenching from 400° C. If the water contains salts (sodium chloride or sulphate) to increase quenching action, crack formation starts at lower quenching temperatures. The number and depth of the cracks increases with increasing mechanical stresses, *i.e.*, with increased quenching temperature. The softening of the material at higher temperatures tends to inhibit the phenomenon, so that above 600° C., crack formation is reduced. The shrinkage stresses are concentrated in the few cracks or even in machining grooves and cause these to widen. The number of cracks is determined by the first quenching and does not alter appreciably during further treatment. The rate of penetration of the cracks is at first rapid and then gradually slows down. At a certain depth a state of equilibrium appears to be reached. Natural or artificial deposits on the cooled side of the steel hinder heat flow and slow down crack formation.—J. P.

Symposium on Fatigue with Emphasis on Statistical Approach—II. (*Amer. Soc. Test. Mat. Spec. Tech. Pub. No.* 137, 1953). This symposium was held at the Society's 55th annual meeting in New York, on June 24, 1952. The papers presented were:

The Statistical Nature of the Fatigue Properties of SAE 4340 Steel Forgings. J. T. Ransome and R. F. Mehl. (3–24).

The Statistical Behavior of Fatigue Properties and the Influence of Metallurgical Factors. E. Epremian and R. F. Mehl. (25–27).

A Statistical Interpretation of the Effect of Understressing on Fatigue Strength. E. Epremian and R. F. Mehl. (58–69).

Fatigue Properties of Large Specimens with Related Size and Statistical Effects. O. J. Horger and H. R. Neifert. (70–91).

Testing Machine for Large Structures—Application of Static and Dynamic Loads. (*Engineering*, 1954, **177**, June 11, 750–753). The article describes a universal testing machine belonging to the 'Association des Industriels de Belgique.' The machine is designed for testing large structures or sub-assemblies for endurance under either static or pulsating loading. The load limit is up to 300 tons vertically and 50 tons in any other direction.—M. D. J. B.

Effect of Shot-Peening Variables and Residual Stresses on Fatigue Life. R. L. Mattson and W. S. Coleman, jun. (*Metal Progress*, 1954, **65**, May, 108–112). The results of a series of tests made on leaf springs are discussed. The specimens were

given various shot-peening treatments and subsequently fatigue tested in groups of four. The influence of shot size, shot velocity, exposure time, and shot-peening whilst under strain were examined.—B. G. B.

The Necessity and the Conditions for Producing Internationally Comparable Rockwell-C Hardness Values. K. Meyer. (*Microtecnic*, 1954, **8**, (3), 138–148). In discussing comparability of Rockwell-C hardness results, the author classifies the possible sources of error, and describes the construction and operation of a standard machine free from mechanical sources of error. The production of suitable test blocks is then outlined, and suggestions are put forward for organizing an international system of comparing hardness values, on a regional basis.—L. D. H.

Methods of Hardness Measurement. (*Usine Nouvelle*, 1954, **10**, June 3, 36). A new Leitz instrument for Vickers and Knoop microhardness tests is described and illustrated. Its advantages are discussed.—T. E. D.

Variation of Hardness of Metals with Grain Size. E. O. Hall. (*Nature*, 1954, **173**, May 15, 948–949). A plot of hardness number against the reciprocal of the square root of the grain diameter yields, for a number of materials, a straight line. This is consistent with the concept of grains yielding at a certain critical stress value built up by dislocations piling up against a grain boundary.—A. G.

The E. Dübi Hardness Characteristic Generalized. A. Collaud. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* MO–2). [In French]. Dübi showed that for a given cast iron there is a linear relation between the values of Brinell hardness and tensile strength obtained from specimens of different diameters cast under the same conditions. Results are presented by the author to show the relations between Brinell hardness, Young's modulus, percentage graphite, tensile strength, and degree of eutectic saturation, for 34 grey irons cast into test-pieces of different diameters. The degree of eutectic saturation is defined after Hanemann as

$$S_e = \frac{C\%}{4{\cdot}23 - 0{\cdot}275 \times P\% - \dfrac{Si\%}{3{\cdot}2}}.$$

Diagrams are presented to show the interrelation between Brinell hardness, S_e, and one or more of the mechanical properties. All the mechanical properties, including bend strength, are plotted on one diagram as functions of Brinell hardness and S_e.—B. C. W.

Wire Hardness Test. L. Small and G. Sevick. (*Metal Progress*, 1954, **65**, Apr., 107–112). A new hardness test which is simple and accurate and can be used with all machines for Rockwell hardness testing is described. It can be used on wire ranging from 0·005 to 0·125 in. dia. and for measuring hardnesses ranging from soft aluminium to tungsten carbide.

Cone Indentation Experiments. D. S. Dugdale. (*J. Mech. Phys. Solids*, 1954, **2**, June, 265–277). In indentation tests with lubricated cones, with included angles ranging from 10° to 140°, on mild steel, copper, and aluminium, the ratio of mean indentation pressure and shear yield stress was found to be the same for the three materials. Minimum possible sizes of indentation specimens, and a suggested volume dependence on the shape of the impression and surface displacements, are calculated and tested.—J. G. W.

Isothermal Temper Embrittlement and the Effect of Hardness on Transition Temperature. B. C. Woodfine. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, May, 532–533). The author shows that the changes in the tough–brittle transition temperature which take place on tempering a low-alloy steel above 650° C. can be explained by the normal processes of tempering, and are not the results of an abnormal embrittling change.—G. F.

Temper-Brittleness of the Medium-C Low Mn–Cr Steels. Z. Takao, K. Takahashi, M. Nishihara, and Y. Tanifuji. (*Tetsu to Hagane*, 1953, **39**, Dec., 1323–1330). [In Japanese]. Steels having varying proportions of phosphorus, molybdenum, and titanium were studied. Low temper brittleness (250–350° C.) is an irreversible phenomenon, caused by decomposition of the martensitic structure; that at high temperature (500–525° C.) is caused by precipitation from the fully tempered structure. The impact transition temperature depends on the ratio Mo%/P%, which should exceed 10 to prevent high-temperature brittleness. No correlation was found between temper brittleness and electrical resistance.

Application of Spheroidal Graphite Cast Iron to Brake Shoes of the Rail Road Tram Car. T. Takase, G. Okamoto, H. Nakamura, T. Terada, and S. Hashiguchi. (*Tetsu to Hagane*, 1954, **40**, Jan., 25–31). [In Japanese]. Practical tests showed that wear loss of brake shoes of spheroidal graphite cast iron was one-third of that of grey iron shoes. The coefficient of friction against the tyres was greater at speeds exceeding 43 km./hr.—K. E. J.

Testing for Wear Resistance. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th Nation Metal Congress*, Oct. 19–23, 1953. Amer. Soc. Metals, 1954, pp. 4–9). The requirements of reliable wear tests are stated, and the statistical evaluation of such tests discussed. Properties governing wear resistance are listed.

Selecting Materials for Wear Resistance. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953: Amer. Soc. Metals, 1954, pp. 10–40). Possible means of protecting metal parts against wear are reviewed. A detailed analysis is given of the various factors in wear, and the selection of resistant materials is discussed in the light of such analyses.—J. E. J.

Martensitic Irons. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953 : Amer. Soc. Metals, 1954, pp. 202–214). The structure of normal chilled cast iron is briefly discussed. A description is given of nickel–chromium, chromium–molybdenum, and chromium–tungsten martensitic irons and their uses together with the modification of their structures and properties by welding.—J. E. J.

Martensitic and Pearlitic Steels. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953. Amer. Soc. Metals : 1954, pp. 231–250). The uses, and effects of variations in composition and heat-treatment on the structure and abrasion resistance of carbon and alloy martensitic and pearlitic steels are described.—J. E. J.

Austenitic Stainless Steels. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953 : Amer. Soc. Metals, 1954, pp. 251–261). The mechanical properties, heat-resistance, and corrosion resistance of the austenitic stainless steels are discussed. The corrosion resistances of types of stainless steels in a wide range of corrosive media are tabulated.—J. E. J.

Austenitic Manganese Steels. H. S. Avery. ('*Surface Protection Against Wear and Corrosion*,' *Educational Lectures at 35th National Metal Congress*, Oct. 19–23, 1953 : Amer. Soc. Metals, 1954, pp. 262–273). The effects of variations in composition on the properties of manganese steels, nickel and manganese steels, and molybdenum and manganese steels are discussed. The inherent disadvantages of this range of steels are described. Methods of hand welding and automatic welding and uses of manganese steel facings are discussed.

An Automatic Plotter for Magnetic Hysteresis Loops. H. McG. Ross. (*Proc. Inst. Elec. Eng.*, 1954, **101**, Part II, Aug., 417–427).

The Ultrasonic Testing of Forging Ingots. R. N. Hafemeister. (*Amer. Soc. Test. Mat. Bull.*, 1954, Apr., 52–55). Tests have proved that it is possible to predetermine the soundness of a forging by determining its condition ultrasonically. This paper describes the technique used to test three forging ingots.—B. G. B.

Portable Ultrasonic Instruments. P. K. Bloch. (*Metal Progress*, 1954, **65**, Mar., 161–166). A short account of this type of instrument and its application in metallurgical testing is presented.—B. G. B.

Symposium on Non-Destructive Testing. (*Amer. Soc. Test. Mat. Spec. Tech. Pub. No.* 145, 1953). This symposium, sponsored by Committee E-7 on Non-Destructive Testing was held at the 55th annual meeting of the Society in New York on June 26, 1952. The papers presented were:

Comparison Radiographs of Welds. A. Gobus and N. A. Kahn. (3–8).

Fluoroscopy and Radiography with Iridium 192. C. Garrett, A. Morrison, and G. Rice. (9–20).

Weld Radiography: A Tentative Method for the Quantitative Evaluation of Defects. O. Masi. (21–30).

Critical Study of Techniques for the Testing of Materials by Ultrasonic Methods. P. G. Bastien. (31–42).

Comparison of Non-Destructive Tests on a Damaged Sternpost. A. de Sterke and H. den Hartog. (43–52).

Progress in the Field of Non-Destructive Testing through the Use of Ultrasonics. W. C. Hitt. (53–75).

The Practical Application of Ultrasonic Non-Destructive Testing. W. A. Felix. (76–88).

Non-Destructive Electronic Sorting of Metals for Physical Properties. F. Förster. (89–98).

The Use and Scope of Iridium 192 for the Radiography of Steel. R. Halmshaw. (*Brit. J. Appl. Phys.*, 1954, **5**, July, 238–243). The use of Ir_{192} for the radiography of steel is discussed and the sensitivities are calculated from absorption coefficients and radiation scatter have been confirmed experimentally. The equivalent X-ray kilovoltage and the range of application of the method as an inspection technique have been estimated.—J. O. L.

The Correlation of the Betatron with Other Forms of Non-Destructive Testing. H. B. Norris. (*Amer. Soc. Test. Mat. Bull.*, 1954, Apr., 56–57). The increased value of magnetic particle, ultrasonic, and low voltage X-ray methods of testing when used in conjunction with the betatron are discussed. The betatron is a high energy X-ray machine whose radiation can easily penetrate 20 in. of steel.—B. G. B.

Materials Testing with Beta-Rays. R. Wideröe. (*Z.V.d.I.*, 1954, **96**, June 1, 450–456). The properties of beta-rays are explained, and their use in the examination of thick-walled steel objects is described, comparisons with X-rays being made. Stereoscopic photography with a double ray is explained, and Brown-Boveri equipment is illustrated.—J. G. W.

Determination of the Best Conditions for Examining Cast Irons by Gamma Radiography by Means of $Cobalt_{60}$. A. Blondel and P. Broquet. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–5). [In French]. The metallurgical, geometrical, and film emulsion factors that can affect the sensitivity of gamma radiography are discussed and results are presented of experiments carried out on different emulsions using a cast-iron specimen and Co_{60} radiation. The sensitivity of the emulsions is compared by plotting density of blackening against specimen thickness for different exposure factors and drawing lines of equal sensitivity. Curves of exposure factor against specimen thickness are also given.

Technical Progress and Economic Aspects of Radiography with Gamma Rays. C. G. Carlström. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* MO–5). [In French]. Recent Swedish work on gamma radiography with Ir_{192} and Co_{60} is reviewed and results given to show the use of the radioactive isotope (Tm_{170}) of the rare-earth thulium as a low-energy source for the radiography of thin (0·5–5 mm.) sections of steel. Definition in radiography is discussed and a mathematical expression derived for sensitivity. The economics of radiography, using either X-ray or X-ray sources, are considered in detail.—B. C. W.

New Device Sorts Metals Accurately, Quickly. E. F. Weller and E. A. Hanysz. (*Iron Age*, 1954, **173**, Mar. 4, 162–165). A thermoelectric metal comparator developed by General Motors Corp. can be used for identifying metal stocks. The instrument uses a pair of metal probes of known alloy which, together with the test sample, make up two thermocouple junctions in series. One probe is heated to a fixed temperature. The differential e.m.f. measured is characteristic of the metal composition. Suitable calibrations can also be made for determining plating thicknesses.—A. M. F.

Rapid Measurements of Thermal Diffusivity. G. E. McIntosh, D. C. Hamilton, and W. L. Sibbitt. (*Trans. Amer. Soc. Mech. Eng.*, 1954, **76**, Apr., 407–410). An apparatus and a technique are described for the determination of the thermal diffusivity of metals by a periodic heat flow method. Measurements with a probable error of $\pm$ 6·8% were made on specimens of Armco iron, titanium, zirconium, and Haynes Stellite 25 at three different temperatures.—D. H.

Apparatus for Determining the Shrinkage and Growth of Metals. I. F. Bol'shakov and T. M. Smirnov. (*Liteinoe Proizvodstvo*, 1954, (3), 16–17). [In Russian]. Two types of apparatus for determining shrinkage and growth of metals are described. In the first, the metal is cast in a special mould forming part of the apparatus, and the variations in the distance between two points on it are followed with the aid of a micrometer device. Examples of results with grey iron, cylinder iron and malleable cast iron, cast at 1290 and 1380° C. are given. In the second, a needle connected to a micrometer and mounted on a stand is embedded in the casting on which the stand is placed. Both types have given satisfactory service.—S. K.

Evaluation of Superheater Materials for High-Temperature Steam. B. Ronay and W. E. Clautice. (*Welding J.*, 1954, **33**, Apr., 199s–206s). An installation for testing alloys exposed to high temperature steam is described. It consists of five test racks: One for a steam temperature of 1100° F.; two racks for steam at 1200° F.; and one rack each for steam at 1350° and 1500° F. respectively.—U. E.

An Experimental Investigation of Res dual Stresses in Hollow Cylinders Due to the Creep Produced by Thermal Stresses. Y. G. Attia, D. Fitzgeorge, and J. A. Pope. (*J. Mech. Phys. Solids*, 1954, **2**, June, 238–258). Residual stresses produced in cast-iron cylinders by the creep-relaxation of thermal stresses were investigated experimentally in connection with failures of combustion chamber parts of large internal combustion engines. Creep in compression at the bore is the predominant cause of residual cooling stresses, which in the extreme case exceed the residual stresses at the outer diameter by a factor of 4½. The creep strains are confined to the inner layers and result in steep residual stress gradients.

Investigation of Heat-Resisting Steel for Gas Turbines (II). S. Koshiba and T. Kuno. (*Tetsu to Hagane*, 1953, **39**, Sept., 991–995). [In Japanese]. In investigations of the ageing of 15%-Ni 20%-Cr austenitic heat-resisting steel, the optimum additions of chromium, tungsten, and molybdenum were determined; appropriate temperatures and times for the solution and ageing treatments were also obtained.—K. E. J.

High Alloy Castings, Heat, Corrosion Resistant. (*Canad. Metals*, 1954, **17**, May 5, 26–27). Types of alloy steels suitable for use at high temperatures are considered. The correct selection of alloys for particular operating conditions are discussed together with some examples of the use of cast heat-resisting alloys.—B. G. B.

Exploratory Creep Tests on Metals of High Melting Point. N. P. Allen and W. E. Carrington. (*J. Inst. Metals*, 1954, **82**, July, 525–533). The strength at 1000° C. of a number of metals, in their purest available form, with melting points above that of iron, was assessed by compression creep tests *in vacuo*, the criterion being the stress needed to give 1% creep strain in 24 hr. Titanium, zirconium, vanadium, niobium, tantalum, chromium, molybdenum, tungsten, platinum, palladium, rhodium, iridium, iron, cobalt, and nickel, and certain sintered carbides were tested. These materials are discussed in the light of their usefulness for high-temperature service.—B. G. B.

Relationships between Long- and Short-Time Creep and Tensile Properties of a Commercial Alloy. A. Graham and K. F. A. Walles. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 105–120). [This issue].

Observations on Elevated-Temperature Tensile Deformation. R. W. Guard, J. H. Keeler, and S. F. Reiter. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Feb., 226–227). For certain tests at elevated temperatures on face-centred cubic, hexagonal close-packed, and body-centred cubic materials, represented by binary alloys of nickel, zirconium, and iron, the authors have found the load-elongation curve to reach maximum load at a relatively small strain, with large amounts of elongation occurring subsequently with continuously decreasing load and, finally, local necking and fracture taking place. This phenomenon is discussed, and the effects of previous strain, strain rate, test temperature, and crystal structure are considered.—G. F.

Problems in Developing Alloy Steels for Aircraft Gas Turbines. F. L. Ver Snyder. (*Iron Steel Eng.*, 1954, **31**, May, 115–123). The author examines the temperatures and stresses to which the different parts of jet engines are submitted. Chemical compositions and rupture properties of different alloys are given together with information showing the variations in hardness with temperature. (13 references).—M. D. J. B.

Hydrogen Embrittlement of Steels. J. T. Brown and W. M. Baldwin, jun. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Feb., 298–303). To characterize more fully the effects of varying strain rate and temperature on the fracture strain of hydrogen-charged steel. The authors have determined the effect of hydrogen on the ductility of 0·20%-C steel at strain rates varying from 0·05 to 19,000 in./in./min., and in the temperature range + 150° to − 320° F. The results are discussed and compared with previous theories of hydrogen embrittlement.—G. F.

Influence of Hydrogen Introduced in the Cold into Cast Irons. P. Bastien, P. Azou, and Christiane Winter. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–37). [In French]. To study the effect of hydrogen on the properties of cast iron it was introduced into specimens of malleable and grey cast iron by direct chemical attack in dilute HCl and by electrolysis in the same solution. The former method was found to be unsuitable as the corrosion of the surface produced effects which masked any influence of the absorbed hydrogen. A malleable iron behaved in a similar manner to steel both in absorption and desorption of hydrogen and in the effect of the gas on the mechanical properties. The capacity of the grey irons to absorb hydrogen reached a maximum and then decreased with increasing time of electrolysis. The effect of hydrogen on the mechanical properties of grey iron and the evolution of the gas from the specimens depended on the structure of the iron.—B. C. W.

Influence of Vanadium and Manganese on the Properties of 5% Chromium Die Steel. N. Yamanaka and K. Kusaka. (*Tetsu to Hagane*, 1953, **39**, Nov., 1271–1277). [In Japanese]. Investigations were made on a 1% C, 5% Cr, 1% Mo steel. The critical point was raised by adding vanadium, and lowered by manganese. Addition of one or the other up to 0·48%, increased the hardenability. Details are given of the best hardening and tempering treatments, and of the volume changes involved. The absorbed energy measured by the static bending test after air-hardening also increased with additions up to 0·48% V.—K. E. J.

New Boron Steel Gives Superior Properties. (*Canad. Metals*, 1954, **17**, May, 52–54). An account of a new boron–molybdenum steel developed by the United Steel Companies and named 'Fortiweld,' is given. The steel has high ductility, fatigue resistance, and tensile strength. It can be welded satisfactorily without any special precautions. It has excellent strength at high temperature and can be readily machined.

German Investigations on Boron Steels. R. Scherer and K. Bungardt. (*Metal Progress*, 1954, **65**, Mar., 101–106, 174–178). The results of a series of tests on boron and non-boron steels are given in detail and discussed. The tests include hardening, heat-treatment and distortion of carburizing steels and the tensile and impact strength of alloy steels.—B. G. B.

Influence of Phosphorus on the Correlation Between Brinell Hardness and the Other Mechanical Properties of Grey Cast Iron. J. Guillamon. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–32). [In French]. Data are presented to show that there is a linear correlation between the Brinell hardness, corrected for phosphorus content, and both the saturation index and the carbon equivalent. The corrected Brinell hardness is given by (measured hardness $- 30 \times$ %P), the saturation index by total C/[4·3 $-$ ⅓(%Si + %P)]and the carbon equivalent by total C + ⅓(%Si + %P). Previous work by the author has shown that other mechanical properties of grey iron are linear functions of the corrected Brinell hardness.—B. C. W.

Literature Survey of High-Strength Steels. (*Welding J.*, 1954, **33**, May, 251s–256s). This survey consists for the most part of a table giving the compositions, mechanical properties, heat-treatment, and weldability of 104 different steels produced by the leading American and European steelmakers.

High-Pressure Vessels Subjected to Static and Dynamic Loads. E. D. Narduzzi and G. Welter. (*Welding J.*, 1954, **33**, May, 230s–238s). An investigation was carried out on the elastic stress–strain relations for external and internal walls for cylindrical shells subjected to internal pressure.

Directionality of Mechanical Properties in Hot-Worked Steels. A. R. Troiano and L. J. Klingler. (*Welding J.*, 1954, **33**, May, 209s–217s). The anisotropy in both cold-worked and hot-worked steel is discussed. The effect of variables such as inclusions, banding, segregates, porosity, and degree of reduction in working was investigated. It was found that inclusions (non-metallic) account for only part of the anisotropy, but that segregation (carbide stringers) and heterogeneous distribution of elements in solid solution accounts for most of the anisotropy.—U. E.

Reports of 22 Technical Committees of the American Society for Metals. (*Metal Progress, Special Issue*, 1954, **66**, July 15, No. 1–A). This special issue constitutes the First Supplement to the 1948 edition of the Metals Handbook. It is confined to reports of 22 technical committees of the American society for Metals. Apart from reports on four non-ferrous metals and their alloys, the remainder are on: Selection of Constructional steels; heat-resisting alloys; nodular cast iron; engine metals and engine wear; effect of stress concentration on design strength; shot peening and other surface working processes; design of ferrous castings; design of brazed joints; carbonitriding; the agitation of quenching media; the cost of heat treating; selection of metal cleaning methods; press forming of sheet steel; machining of steel and cast iron; machining, joining and finishing of powder metal parts; steel melting; selection of methods for non-destructive inspection; and metallography.—R. A. R.

Modern Stainless Steels. R. H. Aborn. (*Metal Progress*, 1954, **65**, June, 115–125). A review of the uses and performance of a wide range of stainless steels is made. Various types of crystalline structure in these steels are considered and pitting corrosion, stress corrosion, intergranular corrosion, and resistance to oxidation are discussed. The mechanical properties of stainless steels and their selection for specific applications are also dealt with.—B. G. B.

Quality of Oxygen-Blown Steel: Properties and Uses of L-D Steels. (Amer. Inst. Min. Met. Eng.: *Iron Coal Trades Rev.*, 1954, **169**, Aug. 20, 473–476). This article deals with the quality of steel produced by the Linz-Donawitz oxygen-blowing process, information being given on the analyses, properties, and applications.—G. F.

Tendency of Automotive Steels and Comparison of Steels Made in Japan, U.S.A. and England. M. Motoyama. (*Tetsu to Hagane*, 1954, **40**, Feb., 142–153). [In Japanese]. Developments in the design of parts and the heat-treatment of steel have reduced the need to use high-alloy steels for automobile parts. Comparisons are made between the grades of steel (including alloy composition, inclusions and grain size) used in the U.S.A., Britain, and Japan.—K. E. J.

Boron Steels. G. Delbart. (*Soc. Roy. Belge. Ing.*, 1953, Dec., 292–310). The manufacture and properties of boron steels are considered in detail. The chemical composition and mechanical properties of boron steels used in America are described. The results of American, British, French, and German research work on these steels are discussed. The properties of a wide range of boron steels are given in tabular form. (35 references).—B. G. B.

Industrial Researches on Boron Steels. G. Delbart and A. Kohn. (Third Report of IRSID Committee for the Study of Boron Steels: *Rev. Mét.*, 1954, **51**, May, 337–363). Trials are reported on five casts of a steel with low contents of nickel, chromium and molybdenum, prepared in an electric or open-hearth furnace with additions of boron in the form of Grainal 79. The results show that, using correct techniques, the industrial production of boron steels presents no special difficulties. The presence of 0·001–0·002% boron considerably increases hardenability and enables the steels so treated to replace those of higher alloy content for certain uses.—A. G.

Effect of Aluminium on the Low Temperature Properties of Relatively High Purity Ferrite. H. T. Green and R. M. Brick. (Trans. Amer. Inst. Min. Met. Eng., 1954, **200**: *J. Met.*, 1954, **6**, Aug., 906–913). True-stress/natural-strain data have been obtained on alloys of pure iron with up to 2·4% Al in the temperature range + 100° to — 185° C. Ductility decreases rapidly as temperature decreases but is still appreciable at — 185° C.; yield and flow stresses increase with decrease in temperature, more rapidly with low aluminium. The effect of temperature and composition on strain hardening are discussed.—G. F.

Rare Earths Counteract Hot-Rolling Defects in Stainless Steel. R. H. Henke and R. A. Lula. (Amer. Inst. Min. Met. Eng. Open-Hearth Conf., 1954: *J. Met.*, 1954, **6**, Aug., 883–888). The authors consider the hot rolling characteristics of martensitic, ferritic, and austenitic stainless steels. Austenitic stainless steels are characterized by two types of hot shortness; that due to duplex structure can be eliminated by correct composition and soaking practice, but the inherent hot shortness of higher alloyed steels is best overcome by addition of an optimum quantity of rare earths, either metallic or oxides. The rare earth practice at the Allegheny Ludlum Steel Corp. and its resulting improvements are described.—G. F.

Precipitation Hardening—How to Work Stainless the Easy Way. A. L. Feild. (*Steel*, 1954, **134**, June 28, 100–102; **135**, July 5, 66–67). Two types of precipitation hardening stainless steels marketed by the Armco Steel Corp. are dealt with, their composition, properties, metallography, methods of working and hardening, and applications being discussed.

The two types are 17-7 PH (17% Cr, 7% Ni, 1% Al) and 17-4 PH (17% Cr, 4% Ni, 4% Cu). They can be easily worked and then hardened by heat-treatment with little scaling and distortion to give high strength and corrosion resistance.

Fortiweld—A New Development in Weldable High-Tensile Steel. H. F. Tremlett. (*Welding and Metal Fab.*, 1954, **22,** June, 221–226). The mechanical properties of Fortiweld steel in the form of plates, sheets, sections, bars and tubes, have been described, and its creep resistance and fatigue properties illustrated. Fortiweld, manufactured by the United Steel Companies, Ltd., is a high-tensile steel of 40 tons ultimate strength and 30 tons yield strength with weldability (freedom from cracking) virtually equivalent to mild steel. Notes on deep pressing and machinability are also included.—U. E.

Study on the Stainless Clad Steel. F. Abe, K. Kimura and T. Saito. (*Tetsu to Hagane*, 1952, **38,** Apr., 214–219; July, 448–453; Dec., 1003–1010). [In Japanese]. Studies on low-carbon steel clad with 18/8 stainless steel showed that casting and assembling were the best mass-production methods. Working and heat-treatment did not affect the adhesion. The tensile strength agreed with the calculated ratio figure, but the endurance limit was higher than calculated; the thermal conductivity was better than that of stainless steel. On heating, particularly above 800° C., deleterious effects occurred, due to diffusion producing an alloy layer of lower binding strength, reduction of corrosion resistance by carbon diffusion, and differential thermal expansion. A very low carbon steel is recommended as the basis metal.—K. E. J.

Researches on the Special Cast Steel. H. Mikashima. (*Tetsu to Hagane*, 1952, **38,** Jan., 2–8; Feb., 86–90; Mar., 122–126; Apr., 196–202; May, 313–321). [In Japanese]. Nickel additions improved in the hardness, impact resistance and other properties of low carbon Cr–Mo steel, but not of all types of cast steel. Manganese additions improved the properties of Cr–Mn and Ni–Cr–Mo steels, the effects reaching their optimum at values of approx. 1·8% and 1% respectively. The hardness of Mn–Si steel was improved by chromium additions up to 1·6%, but the impact resistance was lowered above this figure; additions of tungsten in the range 1·5–2·0% improved the properties. With Cr–Mn–Si steel, a tungsten addition of 0·8% gave excellent resilience. The effects of deoxidation treatments by vanadium, titanium, aluminium, and silicon on the properties of Cr–Mo and Cr–Mn–Si steels are detailed. The effects of various elements on the temper-brittleness of Cr–Mo and Cr–Mn–Si steels are reported. (12 references).–K. E. J.

New Steel Passes Toughness Test. A. G. Gray. (*Steel*, **1954, 135,** July 19, 102–103). A demonstration is described in which four large pressure tanks at − 45° F. were tested to destruction in bursting and impact tests. The material was ' Carillay T-1,' a new extra tough quenched and tempered steel of the United States Steel Co. The composition is C 0·15%; with Mn, Ni, Cr, Mo, B, all less than 1%; the martensite structure is highly developed on water quenching; the yield strength is 90,000 lb./sq. in. Its performance in pressure tanks is outstanding even without stress relieving, but it may be used for any job requiring a structural steel of high strength, great toughness, and good weldability that can be oxy-acetylene cut and cold formed.—D. L. C. P.

Study of the Watch-Spring Material (II). T. Mitsuhashi, M. Ueno, R. Nakagawa, and K. Tsuya. (*Tetsu to Hagane*, 1952, **38,** June, 413–416). [In Japanese]. Six different heat-treatments were employed to examine the effect of cementite size on the mechanical properties of springs. The best results were obtained when the spherical cementite was retained as particles $< 2 \cdot 8\ \mu$ in size.—K. E. J.

Toughness Tests on Lamellar Graphite Cast Irons. G. N. J. Gilbert. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–10). [In French]. Results are presented of impact and slow bend tests on notched and plain bars of pearlitic, acicular, and austenitic irons containing lamellar graphite. In general, the different irons were placed in the same relative order by both Charpy tests and by slow bend tests. Both notched and plain Charpy specimens gave transition curves when tested over a range of temperature. It is concluded that Charpy tests on unnotched bars offer the simplest routine test although Charpy tests on notched bars may be more useful in research work.—B. C. W.

The Influence of the Chemical Composition of Silicate Inclusions on the Properties of Grey Iron. A. Roos. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–8). [In French]. A method for electrolytically separating silicate and oxide inclusions from grey cast iron using a chloride–bromide–citrate solution, is described, and details are given of the methods used to determine the composition of the inclusions and the silica content of the iron. Results obtained on grey cast irons deoxidized with calcium silicide are presented and correlated with the modulus of elasticity.

Elastic Modulus and Structure of Cast Iron. P. Le Rolland and E. Plenard. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* MO–9). [In French]. This is a comprehensive discussion of the elastic properties of grey cast iron and the relationship between these properties and the metallographic structure of the iron.—B. C. W.

The Influence of Copper on Cast Irons Containing Magnesium. R. M. Lamb. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–7). [In French]. Results are presented to show that in magnesium-treated cast iron, copper suppresses the formation of spheroidal graphite and also stabilizes the pearlite. The graphite present as spheroids decreases to zero at 3% Cu, then increases to about 70% at 4–5% Cu, and finally decreases again at higher copper contents. Magnesium appears to lower the solubility of copper in cast iron and the adverse effect of copper on the form of the graphite is associated with the appearance of a copper-rich phase in the microstructure. Additions of over 4% Cu completely suppress the presence of ferrite. The addition of 0·02% mischmetall neutralizes the effect of copper on the formation of spheroidal graphite but does not affect either the solubility of copper or the stability of the pearlite. Alloys of Cu–Mg are less effective in introducing magnesium into the cast iron than Ni–Mg alloys.—B. C. W.

The Effect of Copper on Some Properties of Grey Iron. J. Foulon and A. De Sy. (*Fonderie*, 1954, Jan., 3755–3774: *International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–16). [In French]. The published work on copper in cast iron is briefly reviewed and results are presented of tests carried out on as-cast grey irons containing different amounts of carbon, silicon, and manganese, with up to 5% Cu. In all cases increasing the copper content increased the hardness but decreased the impact value. The tensile strength increased up to 3% Cu and then decreased. The results are discussed in relation to the effects of copper on the microstructure and its solubility in α- and γ-iron.—B. C. W.

Report on Whiteheart Malleable Irons Containing Molybdenum or Manganese and Chromium. F. Roll. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–20). The properties of whiteheart malleable irons containing either 0·1–0·15% Mo, or 1% Mn and 0·2% Cr, are briefly reviewed. These alloyed irons have better mechanical properties and improved wear resistance in service.—B. C. W.

Specifications and Tests for Spheroidal Graphite Cast Irons. M. Ballay and J. Grilliat. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–21). [In French]. American and French specifications for spheroidal graphite cast irons are briefly reviewed and results are given of tensile tests made on different irons using test-pieces of different dimensions. A bend test is described for spheroidal graphite irons and data obtained with it are compared with the results of tensile tests. The bend test has the advantage of using a very small specimen.—B. C. W.

Influence of Various Elements on Mechanical Properties of Cast Iron Having Eutectic Graphite Structure Produced by Treating Molten Cast Iron with Slag Containing Titanium Oxide (S-H Cast Iron). (II). H. Sawamura, Y. Hotta, and T. Oka. (*Tetsu to Hagane*, 1953, **39,** Sept., 980–983). [In Japanese]. Manganese had a marked effect in improving the tensile strength, transverse strength and hardness of the sand-cast iron, but copper had a considerably smaller effect.

The Ductility of Whiteheart Malleable Iron. G. N. J. Gilbert. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5,** Feb., 160–172). Whiteheart malleable iron of balanced sulphur content having a black fracture is shown to have impact properties at room temperature superior to those of irons having the conventional whiteheart fracture. Impact tests were carried out on four whiteheart malleable irons; two irons annealed in ore and two by the gaseous process. The tests were conducted at temperatures between − 180° and + 200° C. for each material. The influence of structure and composition on the

impact transition temperature and impact values are discussed.—B. G. B.

The Magnetic and Electrical Properties of Cast Iron. M. H. Hillman. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5**, Apr., 188–248). The results are given of an investigation of the magnetic and electrical properties of a large number of different types of cast irons with varying combinations of graphite and matrix structure. The remanence, coercive force, hysteresis loss, maximum permeability, and specific resistance were determined for 24 specimens. The microstructures are illustrated in each case, together with the appropriate magnetization curve and hysteresis loop. Details of composition as determined by chemical and spectrographic analysis are also included.—B. G. B.

Temper Embrittlement in Blackheart Malleable and Nodular Cast Iron. G. N. J. Gilbert. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1954, **5**, Apr., 249–263). The results of a detailed investigation of temper embrittlement in these two types of iron are given and discussed. The conditions of formation of brittle structures and suitable methods of reducing the embrittlement are considered.—B. G. B.

On the Relationship Between the Microstructure and the Modulus of Elasticity of Cast Iron. M. Okamoto and T. Tottori. (*Tetsu to Hagane*, 1952, **38**, May, 306–312). [In Japanese]. The modulus of elasticity is mainly affected by the graphite present, increasing as the form varies from flaky to spheroidal. Increase in quantity of graphite of the same form lowers the modulus. Changes due to the number of graphite nodules, matrix structure or silicon concentration, are small compared with those due to the form or quantity of graphite. With rise in temperature, the modulus decreases slowly at first, then rapidly above 300° C.—K. E. J.

Decrease of Density During Plastic Deformation of Nodular Cast Iron. W. R. Clough and M. E. Shank. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200** : *J. Met.*, 1954, **6**, Sept., 1093–1094). The authors have investigated the pronounced decrease in density which occurs when nodular cast iron is subjected to stress. From microscopical examinations they conclude that the density decrease is due to the opening up of holes around the graphite nodules, and that initial separation occurs in a thin external ring of the nodule itself.—G. F.

Hardness and Tensile Strength of Annealed Spheroidal Graphite Cast Iron. A. L. Carr. (*Iron Coal Trades Rev.*, 1954, **169**, Aug. 27, 507–510). The author has determined the effects of silicon, manganese, nickel, and phosphorus, and of the proportion of residual pearlite, on the mechanical properties of annealed laboratory-cast spheroidal graphite cast iron. The results are presented, and the relationships between yield point, tensile strength, and hardness of the material are discussed.—G. F.

National Physical Laboratory Extension. (*Metallurgia*, 1954, **49**, May, 257–260). A description of the new laboratory of the metallurgical division of the National Physical Laboratory which was opened in April, 1954, is given.

METALLOGRAPHY

A Laboratory Arc-Melting Furnace. M. L. Nielsen and I. B. Johns. (*Rev. Sci. Instruments*, 1954, **25**, June, 596–598). A small arc furnace is described for melting metals in an atmosphere of argon.—L. D. H.

Experience with Electron Microscopy at Metallografiska Institutet, Stockholm. S. Modin. (*Jernkontorets Ann.*, 1954, **138**, (8), 479–490). [In Swedish]. A description of the Institute's large RCA–EMU 2D electron microscope includes brief notes on certain factors limiting its resolving power. Stereo and diffraction applications are touched upon and a simple method given for increasing the contrast in electron-optical photographs of replicas. A detailed account is given of replica technique employed.—G. G. K.

Methods of Investigating Metallic Structures with the Aid of the Electron Microscope. D. S. Shraiber, E. K. Molchanova, and N. S. Shcherbakova. (*Zavodskaya Laboratoriya*, 1950, **16**, (11), 1321–1330). [In Russian]. This is a general review of the application of electron-microscopic methods to metallography. After enumerating the metallographic applications, an account is given of electrolytic polishing and etching, and of the preparation of surface replicas. Some photomicrographs at 15,000 diameters are shown.—S. K.

Use of Large Magnifications for Studying the Structure of Steel. A. P. Gulyaev, A. G. Zavrazhin, and I. A. Strel'nikov. (*Zavodskaya Laboratoriya*, 1950, **16**, (11), 1335–1336). [In Russian]. The structures of an iron and two steels (1·0 and 1·54% C, respectively) were examined at magnifications of 7000 diameters with the aid of an electron microscope. Photomicrographs are presented and discussed. Ferrite grains were found to be differently inclined to the plane of the section, and high magnifications are specially recommended for the study of grain boundaries. Martensite needles were divided into two main parts by deep cracks. High magnification yielded no new data on the structure of lamellar pearlite.—S. K.

A Polishing Machine for Micro Specimens. L. Ekbom. (*Jernkontorets Ann.*, 1954, **138**, (8), 490–494). [In Swedish]. In an attempt to reduce the excessive distance between the specimen holder and the polishing disc, which with the friction due to polishing, tends to twist the specimen, a new machine has been constructed which allows the specimen to rest freely on the table, its movement being guided by a ring attached to the machine arm. The gap between ring and table is small, thus obviating any tendency to tip. Three or more specimens can be polished simultaneously, weight-loaded at will, or removed without stopping the polisher. The polishing time is 10–15 min.—G. G. K.

Contribution to the Theory of Electrolytic Polishing. E. Knuth-Winterfeldt. (*Arch. Eisenhüttenwesen*, 1954, **25**, July-Aug., 393–395). The theory of electrolytic polishing has been extended by a study of heat evolution during the polishing procedure. An anodic film with a high temperature maximum within it is favourable to good polishing, whereas when the temperature maximum is at the metal surface, conditions are not so suitable. The occurrence of pockets of heat, *i.e.*, regions within the film of marked temperature difference, appears to be of great importance for electrolytic polishing. The shape of the current-density/potential curve for the anodic treatment of copper in phosphoric acid is discussed in detail with reference to various theories.—J. P.

Measurement of Grain Size in Metals and Alloys. E. Dupuy. (*Métaux-Corrosion-Indust.*, 1954, **29**, June, 225–241). The measurement of grain size is considered in detail. Different methods, including planimetry, counting the grains in a given area, and comparison with standard samples are explained. The theory, practice, and precision of the different methods are discussed and the choice of method to be used in particular conditions is indicated. The American A.S.T.M., Jernkontoret (JKN) and French (AFNOR) standard methods are described and the relationship between them is discussed. (23 references).

Defects in Steel, Methods of Detecting Them. (*Tech. Indust.*, 1954, **32**, Mar., 192–194). [In Spanish]. A brief description of the preparation of surfaces for metallographic examination in order to detect segregations and other defects is given. Mention is also made of inclusion counting and various methods of crack detection.—C. B. L.

On the Nucleation of Pearlite. M. E. Nicholson. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**; *J. Met.*, 1954, **6**, Sept., 1071–1074). The author reviews and analyses evidence on the nucleation of pearlite and puts forward a theory that it may be initiated by nucleation of either cementite or ferrite from austenite, depending upon composition and temperature. This two-mode model appears to be in accord with all observed structural relations and rates of pearlite nucleation, and explains the loss of hardenability effect with increasing carbon in boron-treated steels.

The Application of Colour Metallography to the Study of Spheroidal Graphite Cast Iron. A. Scortecci and C. Durand. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* MO–4). [In French]. (*Fonderia Ital.*, 1954, **3**, Apr., 143–147). [In Italian]. Colour photomicrographs from heat-tinted specimens of lamellar graphite, blackheart malleable and various types of spheroidal graphite cast iron are presented and discussed. Among the points observed was that the ferrite 'halos' surrounding the graphite spheroids were in general of a uniform colour implying a uniform rate of oxidation. Theories to account for this are discussed and it is suggested that the rate of oxidation of the surface is principally determined by composition, which masks any effect produced by orientation changes between grains.

Studies on the High-Speed Tools (XVI). H. Hotta and I. Tatsukawa. (*Tetsu to Hagane*, 1953, **39**, Oct., 1177–1182). [In Japanese]. Results are reported for a high-speed steel of composition C 0·65%, W 15·92%, Cr 4·30%, V 0·76%, and

Co 2·64%. Austempering at 600° C. led to bainite production on cooling, with little softening on tempering at 300–400° C. After austempering at 250–300° C. the final structure contained appreciable amounts of retained austenite, which decreased the room temperature hardness and caused secondary hardening on tempering. The formation of bainite on austempering increased the stability of retained austenite. (22 references).—K. E. J.

On the Growth of Ductile Cast Iron. K. Nagaoka. (*Tetsu to Hagane*, 1953, **39**, Nov., 1250–1258). [In Japanese]. The growth of ductile and other cast irons was studied in air and vacuum, during repeated cyclic heating, by dilatometric and microscopic means. Its growth was less than those of other types of iron; it increased with temperature, but oxidation effects or decrease of strength were not apparent, nor were the Ar_1 expansion or Ac_1 contraction. The growth is attributed to the locational irreversibility of graphite formation, varying with the mechanism of ferrite formation through the critical range and oxidation effects. (11 references).

The Sigma Phase in Binary Alloys. P. Greenfield and P. A. Beck. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Feb., 253–257). Since the experimental data for the composition ranges of several of the known σ phases is rather incomplete, the authors have made new experimental observations for a number of binary systems, including V–Fe, Cr–Fe, and Mo–Fe. Two alternative criteria for describing the composition of σ phases are discussed, and it is felt that the relative atomic sizes of the alloy components significantly affect σ formation.—G. F.

Sigma—an Unwanted Constituent in Stainless Weld Metal. L. K. Poole. (*Metal Progress*, 1954, **65**, June, 108–112). The formation, control, and identification of the sigma phase, which is formed in stainless steels and weld metals of austenitic structure at elevated temperatures, are discussed. The mechanism and rate of sigma formation are also considered.

Sigma Phase—A Review. A. J. Lena. (*Metal Progress*, 1954, **66**, July, 86–90). Sigma phase formation in ferritic and austenitic steels is considered and it is shown that a large number of cast and wrought alloys are capable of developing this phase. These include essentially all the Amer. Iron Steel Inst. steels with $> 17\%$ Cr, as well as the newer austenitic chromium–manganese steels.—B. G. B.

Carbides in 310 Stainless Steel. H. Brown. (*Steel*, 1954, **135**, July 19, 106–108; July 26, 92–94). This steel is a 25/20 Cr–Ni heat and corrosion resistant alloy and the author considers the complex situation wherein chromium carbide precipitation occurs round grain boundaries. This sometimes leaves the steel sensitive to corrosive attack; it is suggested that this is because chromium adjacent to the grain boundaries may be depleted.—D. L. C. P.

Metallography of Gray Cast Iron, Blackheart Malleable Iron, Spheroidal Graphite Iron. A. R. Cluett. (Inst. Brit. Found.: *Engineer Foundryman*, 1953, **18**, Sept., 88–98). An elementary account is given of the metallography of cast irons.

Nucleation and Growth of Graphite in Steel. F. Brown. (*Welding J.*, 1954, **33**, June, 257s–261s). Measurements on the nucleation and growth of graphite in steel are reported.

Structure of Graphite Spherulite. M. N. Parthasarathi and B. R. Nijhawan. (*Trans. Indian Inst. Met.*, 1952, **6**, 150–169). An attempt is made to explain the structure of the graphite nodules and to elucidate some of the outstanding controversial points. The significance of the central white spot and the radial type of nodule is discussed in the light of experimental data obtained by studying four nodules cut across different parallel sections.—D. H.

Impurities as Hindrances to Grain Boundary Migration. V. G. Paranjpe. (*Trans. Indian Inst. Met.*, 1952, **6**, 203–218). Detailed mechanisms of the hindrance offered by inclusions or impurity concentrations to grain boundary migration are outlined and supporting evidence is cited. A formula is derived for the hindrance offered by a row of inclusions. Their cumulative effect is considered in order to explain the temperature dependence of the grain growth exponent and the heat of activation for grain growth. Secondary crystallization takes place when the hindrances are eliminated or surmounted.—D. H.

A Split-Image Film Comparator for X-Ray Diffraction Powder Patterns. J. T. Quan. (*Rev. Sci. Instruments*, 1954, **25**, Apr., 399–401).

A Precision Micro-Thermostat for Crystal X-Ray Studies. P. J. A. McKeown. (*J. Sci. Instruments*, 1954, **31**, Aug., 271–273).

Adaptation of a Geiger-Counter X-Ray Diffractometer for High-Temperature Investigations. P. Chiotti. (*Rev. Sci. Instruments*, 1954, **25**, July, 683–688). A specimen holder and furnace are described for use with a high angle goniometer Geiger counter X-ray diffractometer. The equipment has been used satisfactorily up to 1500° C., although patterns have been obtained up to 1650° C. Temperatures up to about 900° C. are measured by a thermocouple inserted in the specimen; for higher temperatures an optical pyrometer is used.—L. D. H.

Standard X-Ray Diffraction Powder Patterns. (*U.S. National Bureau of Standards*, 1954, **3**, *Circular* 539).

A Microscopical Examination of Samples of Iron Containing Silicon Inclusions. R. E. Lismer and F. B. Pickering. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 159–162). [This issue].

The Domain Structure of Ferromagnetic Metals. L. F. Bates. (*J. Inst. Metals*, 1954, **82**, May, 417–425). A brief outline is given of the domain concept as applied to ferromagnetic materials, metals and their alloys. A description of the technique used to map out the intersections of domain boundaries, is given. The value of the results obtained is explained.—B. G. B.

Re-Examination of the Symmetries of Iron and Nickel by the Powder Method. F. W. von Batchelder and R. F. Raeuchle. (*Acta Crystallographica*, 1954, **7**, May, 464). Lattice constants measurements reported here do not support Kochanouska's conclusions that deviation from cubic symmetry develops in iron above 250° C., or that the deviation could be observed by X-ray techniques.—T. E. D.

The Nature of Dislocations in Ideal Single Crystals. J. S. Koehler and F. Seitz. (*Amer. Inst. Min. Met. Eng. Seminar on "Dislocations in Metals,"* Oct., 1951, 1–36). The need for dislocations to account for the great variety of plastic flow phenomena is discussed and the geometry, stress fields, energies and interactions of dislocations are treated in considerable detail. The interrelation between dislocations and other crystal imperfections is also dealt with.—A. G.

Rôle of Dislocations in Crystal Growth and Grain Boundary Phenomena. W. T. Read, jun., and W. Shockley. (*Amer. Inst. Min. Met. Eng. Seminar on "Dislocations in Metals,"* Oct., 1951, 37–68). Recent developments in dislocation theory are reviewed with the emphasis on those aspects where definite predictions have been subsequently verified by experiment. The problems considered are the Frank theory of crystal growth, low angle grain-boundary energies, and the movement of these boundaries under stress.—A. G.

Method of Using a Fine-Focus X-Ray Tube for Examining the Surface of Single Crystals. L. G. Schulz. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Sept., 1082–1083). Details are given of a method of examination of the surfaces of single crystals. White radiation from a fine-focus X-ray tube is used to produce an image of a relatively large area of the crystal surface.—G. F.

Mechanism for the Original of Recrystallization Nuclei. J. P. Nielsen. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Sept., 1084–1088). Based on grain-boundary free-energy considerations, the author proposes a 'geometrical coalescence' mechanism as a logical origin for recrystallization nuclei, particularly for secondary recrystallization.—G. F.

Examination of the Mechanical Properties of Steels by Magnetic Methods. P. E. Lagasse. (*Rev. Univ. Min.*, 1954, 9th series, **10**, Sept., 608–616). The magnetic properties of steels are related to their crystalline structure and this is used to study the changes which occur during heat-treatment. The principles of the method are described and examples of its application are given. (15 references).

The Climb of Edge Dislocations in Face-Centred Cubic Crystals. R. S. Barnes. (*Acta Met.*, 1954, **2**, May, 380–385). Conditions for dislocations to move out of their slip plane (or 'climb') are analysed. The effect is thought to be important chiefly at high temperatures or during particle bombardment.—A. D. H.

On the Theory of Secondary Recrystallization Textures in Face-Centred Cubic Metals. C. G. Dunn. (*Acta Met.*, 1954, **2**, May, 386–393). An orientated-nucleation growth-selectivity theory is discussed. The theory is used to explain previously reported results.—A. D. H.

The Carbides Formed in the Bainite Transformation in Steels. J. Pomey. (*Génie Civil*, 1954, **131**, Aug. 15, 314–315). The difference between the ζ orthorhombic carbide formed in pure, 1·4%-Si, or 2%-Ni eutectoid carbon steels by the isothermal transformation of austenite to bainite at 300° C., and the ε hexagonal percarbide formed in a similar steel with isothermal bainitic temper is mentioned. Thermomagnetic differential curves indicative of the different carbides are shown.—T. E. D.

Effect of Tempering on the Hardness of Retained Austenite. P. Stark and B. S. Lement. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Sept., 1074–1075). The authors have investigated the reported hardening of retained austenite in 1·7%-C steels curved by subsequent tempering, using pure Fe–C alloys. They conclude that the reported increase is not due to hardening of the austenite only, but to the formation of surface martensite.—G. F.

Dislocations and Mechanical Properties. E. Orowan. (*Amer. Inst. Min. Met. Eng. Seminar on "Dislocations in Metals,"* Oct., 1951, 69–195). Early attempts to account for the discrepancy between observed and calculated yield values are discussed and the considerations which led to attributing this discrepancy to the existence of dislocations in crystalline materials are described. The mechanisms of origin of dislocations are outlined and the application of dislocation theory to a wide range of mechanical properties including the yield point, creep, precipitation and strain hardening, ductile fracture, and fatigue is treated in detail.—A. G.

Diffuse X-Ray Scattering and the Physical Properties of Crystals. W. A. Wooster. (*Brit. J. Appl. Phys.*, 1954, **5**, July, 231–237). Ionization and photographic studies of X-ray reflections can be related to thermal agitation of atoms and the elastic constants of cubic crystals. Interphase precipitates have characteristic reflection patterns.—J. O. L.

Primary and Secondary Recrystallization in Cold Rolled Bicrystals of Silicon Iron. C. G. Dunn. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, May, 549–550). The author traces the origin of secondary recrystallization grains produced on annealing cold-rolled bicrystal specimens of 3·25% Si iron. The results indicate that a large primary grain can grow into a secondary grain.—G. F.

Deformation of Iron Crystals by Unidirectional Abrasion. R. P. Agarwala and H. Wilman. (*J. Iron Steel Inst.*, 1955, **179**, Feb., 124–131). [This issue].

The Nature of Lamellar Graphite. D. P. Ivanov. (*Liteĭnoe Proizvodstvo*, 1954, (3), 18–24). [In Russian]. An investigation of graphite structure, especially lamellar, in cast iron is reported. Photomicrographs of polished sections from cast irons solidifying under different conditions are shown. Experiments showed that the initial structure of graphite in Fe–C alloys is a compact arrangement of hexagonal blocks of plates. All other structural forms of graphite are derived from this, their specific forms depending on the conditions of crystallization.—S. K.

Graphitization of Pure Iron–Carbon Alloys. K. P. Bunin and T. M. Shpak. (*Liteĭnoe Proizvodstvo*, 1954, (3), 26–27). [In Russian]. The object of the investigation of the graphitization of relatively pure Fe–C alloys described was to find the rate-controlling factor in this process. A hypoeutectoid white iron (3·52% C, 0·02% Mn, 0·08% Si, 0·01% S) was prepared from electrolytic iron and electrode graphite. No graphite could be detected in the alloy as cast. Specimens were heated for various times under carefully controlled conditions, the graphitization process being followed metallographically. The results obtained show that, except for graphitization in pores, the rate is controlled by the displacement of iron atoms and not by diffusion of carbon.—S. K.

Tests to Elucidate the Transformation Behaviour of a Special Carbide-Forming Chromium Steel. F. Wever and W. Koch. (*Stahl u. Eisen*, 1954, **74**, July 29, 989–1000). The isothermal transformation of a 0·43%-C 3·5%-Cr steel in the pearlite and intermediate ranges has been studied and the carbides separated electrolytically and examined by microchemical, X-ray, and electron microscopic methods. The composition, crystal form, and lattice parameters of the carbides formed in the two ranges and their change with holding time are described.—J. P.

The Characteristics of the Martensite Transformation. D. Hull. (*Bull. Inst. Metals*, 1954, **2**, July, 134–139). After a brief historical survey, an attempt is made to define the martensite transformation in terms of its diffusionless characteristics. The validity of this definition is discussed in relation to the known characteristics of the transformation under four headings: (1) Diffusionless characteristics; (2) crystallographic characteristics; (3) strain characteristics; and (4) other characteristics. (28 references).—B. G. B.

Calculation of Thermodynamic Data in Metallurgical Processes. A. Krupkowski. (*Archiwum Gornictwa i Hutnictwa*, 1953, **1**, (2), 107–129). [In Polish]. The difficulties in calculating the free energy change and equilibrium constants for various reactions when phase changes occur in some of the reacting components are discussed. It is claimed that the introduction of standard thermodynamic data for supercooled substances by A. Krupkowski and W. Ptak removed the usual difficulties in the calculation of free energy changes and equilibrium constants. Using these data the free energy change in the temperature range 1800–2000° K. of a number of reactions are calculated.—V. G.

Application of Transformation Diagrams to Special Problems in the Fabrication of Highly Stressed Welded Structures. F. Nehl and A. Rose. (*Stahl u. Eisen*, 1954, **74**, Aug. 12, 1054–1061). The development of weldable heavy structural steels towards increased mechanical properties is discussed with particular reference to the increased tendency to hardening crack formation as the alloy content is raised. The possibilities of improving mechanical properties by alloying and heat-treatment and also the danger of excessive martensite formation during the rapid cooling of the transition zone of the weld are illustrated by means of continuous time-temperature-transformation diagrams. These cooling processes have been studied by measurements and from the structure and the hardness, with the help of continuous time-temperature diagrams. Cu–Ni–Mo weldable steels (0·15% C, 1% Mn, 1% Cu, 1% Ni, 0·35% Mo, 0·4% Cr) have high tensile strength and low tendency to hardening cracks on welding. The reason for this is that these steels transform in the intermediate range over a wide temperature zone, from the slow air cooling of thick plates to the rapid cooling in the transition zone of welds.—J. P.

Isothermal Transformations and Heat-Treatments of Copper–Chromium and Nickel–Chromium Cast Irons. J. Van Eeghem, J. Vidts, and A. de Sy. (*International Foundry Congress*, Paris, Sept. 19–27, 1953, *Paper No.* CI–17). [In French]. The isothermal transformation diagrams which were determined dilatometrically and metallographically for two copper–chromium and two nickel–chromium cast irons, are presented together with micrographs of the transformation products. The relative effects of copper and nickel on the transformations are discussed. In all alloys pearlite was not nucleated by the graphite flakes but both upper and lower bainite were.—B. C. W.

Constitution of Iron–Boron Alloys in the Low Boron Range. M. E. Nicholson. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, Feb., 185–190). The author has determined the solid solubility of boron in iron by saturating vacuum-melted high-purity iron with respect to Fe_2B in the temperature range 870–1135° C. The maximum solubility in α-iron is 0·002% B and in γ-iron varies from 0·001% B at 915° C. to 0·020% B at 1165° C. The eutectic and peritectoid temperatures are respectively 1165° and 911° C. and the position of the γ-iron solidus approximates to 0·020% B. A constitutional diagram based on the investigation is given.

A Rationalization of the Oxygen Solid Solubility in Some Transition Metals. A. V. Seybolt and R. L. Fullman. (*Trans. Amer. Inst. Min. Met. Eng.*, 1954, **200**: *J. Met.*, 1954, **6**, May, 548–549). From available data, the authors have examined the possibilities of a simple relationship occurring between oxygen solid solubility and the properties of a number of metal solvents, including α-iron. The logarithm of the maximum solubility is found to bear an approximately linear relationship to the ratio of elastic modulus to the volume of an octahedral interstitial space.—G. F.

The Solid Solubility of Phosphorus in Iron. H. Schrader and B. N. Bose. (*Trans. Indian Inst. Met.*, 1952, **6**, 104–136: *Eastern Met. Rev.*, 1953, **6**, Apr. 6, 407–411). The effect of phosphorus on the mechanical properties of steel is discussed. The increase in yield strength with a decrease in toughness together with an ageing effect at room temperature indicates that the embrittlement of steels high in phosphorus should be due to a precipitation effect rather than to grain-coarsening.

This is supported by the reported tendency for carbon to enlarge the γ loop and accelerate ageing. Ageing is not understandable from the existing Fe–P binary diagram and may be explained on the basis of phosphorus segregation. Following the lead of Zener, a thermodynamical computation is presented for the calculation of phase boundary relationships. Information is collected and evaluated on the effect of carbon, nickel, chromium, and manganese of the solubility of phosphorus in iron.—D. H.

CORROSION

A Year's Work on Corrosion at the Chemical Research Laboratory. (*Corrosion Technology*, 1954, **1**, June, 107–110). A description is given of the current research being carried out by the Corrosion of Metals Group of the Chemical Research Laboratory, Teddington, Middlesex.—L. E. W.

National Convention on Combatting Corrosion. (*Met. Ital.*, 1954, **46**, Supplement to No. 5). The whole of this issue is devoted to reporting this symposium on corrosion, held in Milan on October 8–10, 1953, under the auspices of the *Consiglio Nazionale delle Ricerche*, and organized by the *Centro Elettrochimica*, *Associazione Italiana di Metallurgia*. The papers presented, which are all in Italian, were:

The Phenomena of the Passivation of Metals and Alloys. R. Piontelli. (5–28).

The Passivity of Zinc in Relation to the Conditions of Aeration. G. Bianchi. (29–32).

The Phenomena of Passivity in Stainless Steels. A. Ferri. (33).

Conditions of Passivity Produced by Oxidizing Agents on Ordinary and Special Steels. A. Indelli. (34–35).

Notes on the Use of Metallic Materials and on Corrosion in the Chemical Industry. G. Pastonesi. (37–41).

The Influence of Gases in Solution on the Pitting of Stainless Steels. C. Bighi. (42–43).

Corrosion Problems and Measures of Protection in a Petrol Refinery. P. Daino. (47–50).

A Review of Materials Available Today for Resisting Corrosion Phenomena. P. Negretto. (51–53).

Drinking Waters and Their Corrosive Action on Ferrous Materials. R. Sandrinelli. (54–58).

Anticorrosive Protection by Means of Inhibiting Substances. L. Cavallaro. (59–63).

Contribution to the Study of Inorganic Inhibitors. U. Bertocci. (64–66).

Anticorrosive Protection by Phosphatization. P. De Cerna. (67).

The Use of Inhibitors in the Cleaning of Castings. B. Domenicali. (68).

The Use of Inhibitors for the Chemical Descaling of Boilers. B. Domenicali. (69).

Research on Activating and Wetting Inhibitors for Cleaning, and on Cold Phosphatization. L. Felloni. (70–72).

Adsorption and the Action of Inhibitors for Cleaning. A. Indelli and G. P. Bolognesi. (73–74).

Research on Inhibitors in the Vapour Phase. G. Mantovani. (75–76).

Practical Considerations Relating to the Theoretical Conceptions on the Anticorrosive Protection of Manufactured Goods. F. Schieroni. (77, 95).

Corrosion and Protection of Buried Metallic Materials. M. Jacopetti. (79–87).

The Use of Tin for Protection Against Corrosion. S. C. Britton. (89–90).

Cases of Corrosion in Panel Radiator Heating Pipes. E. Donati and G. Beretta. (91–95).

Cathodic Protection of Buried Ducts and Metal Structures. E. Gerosa. (96–97).

Economic Aspects of Cathodic Protection. T. Nanni and A. Compostella. (98–100).

Contribution to the Study of the Oxidation of Heated Metals and Alloys. F. De Carli and P. Spinedi. (101–107).

Methods of Testing and Controlling the Corrosion of Metals. G. Bianchi. (109–112).

Ceramic Refractory Covering for the Protection of Metals at High Temperatures. E. Crepaz. (113–116).

Methods of Testing the Action of Pigments on Varnishes. C. Bighi and G. Mantovani. (117–118).

Determining Inhibitor Power by Means of Polarization Curves. G. Bombara. (119–120).

The Use of Electrochemical Methods for Testing Materials. L. Cavallero and A. Indelli. (120–121).

Corrosion Research at the Istituto Sperimentale dei Metalli Leggeri. G. Luft. (122–124).

Analysis of Metallic Sounds for Measuring Intercrystalline Corrosion in 18/8 Stainless Steels. O. Masi and A. Ferri. (125–127).

Continuous Polarographic Control of the Kinetics of Corrosion. L. Riccoboni, P. Papoff, and V. Genta. (129).

Analytical Methods Applied to the Study of Drinking Waters and Soils to Determine Their Corrosive Activity on Ferrous Materials. R. Sandrinelli, P. Sbrascia, and A. Andreini. (130–132).

Laboratory Examination of the Corrosive Activity of Soils on Steel and Cast Iron Pipes. R. Sandrinelli. (132–135).

Research Techniques and Information Available to Combat Corrosion. M. Pourbaix. (136–140).

Italian Standardization in the Field of Corrosion of Metallic Materials. (141).

The Electrolysis Division of the Società Nazionale Metanodotti. A. Grandi. (142–144).

The Corrosion Committee of the Società Montecatini. Problems Handled by the Committee. E. Hugony. (145–146).

Organization and Activities of the Corrosion Research Division of the Breda Scientific Research Institute. L. Matteoli. (147–148).

Electrochemical De-Rusting. (*Corrosion Technology*, 1954, **1**, June, 116–117). A description is given of an electrochemical process for derusting corroded equipment and plant. The process removes rust and other surface deposits but does not render the surface corrosion-resistant. This method of derusting has been operated successfully by Derustit (Great Britain) Ltd. The complete process consists of five stages: Electrolytic degreasing, cold water rinse, electrolytic derusting, hot water rinse, and treatment with de-watering oil or other form of protection.—L. E. W.

Cathodic Protection. L. L. Shreir. (*Chemical and Process Engineering, Corrosion Technology*, 1954, **1**, May, 63–65, 78; June, 103–106). Underground corrosion with particular reference to sulphate-reducing bacteria, chemical nature of the soil, stray currents, concentration cells, and dissimilar metals in contact, is discussed and the theoretical principles involved in cathodic protection are dealt with. Methods of cathodic protection are described. (37 references).—L. E. W.

Minimizing Damage to Adjacent Structures in Urban Cathodic Protection Installations. D. T. Rosselle. (*Corrosion*, 1954, **10**, June, 192–194). Electrolytic damage to underground structures adjacent to other structures receiving cathodic protection can be avoided by correct bonding. Bonds should be carefully planned and calibrated and provided with test leads. An example is given where cross-bonding was inadequate to protect a telephone cable adjacent to a cathodically protected pipe. Distributing anodes alongside the protected structure and the use of many small rectifiers will reduce damage to neighbouring plant.—J. F. S.

Cathodic Protection in Florida Power and Light Company's Generating Plants. J. B. Prime, jun. (*Corrosion*, 1954, **10**, May, 165–168). Experience with the application of cathodic protection to the intake structures and sea-water cooled condensers of an electricity company is described.—J. F. S.

Corrosion and Design. L. L. Shrier. (*Chemical and Process Engineering, Corrosion Technology*, 1954, **1**, Mar., 6–9; Apr., 39–42). The author applies the theoretical principles of corrosion to practical problems and considers in particular how corrosion can frequently be avoided in the earliest stages by attention to design. No specific process is considered but rather the effect of the metal, method of construction, shape, velocity of flow of the liquid, dissimilar metals in contact, on the possibility of corrosion occurring. Designing to avoid corrosion should be followed by pilot plant testing. (34 references).—L. E. W.

Corrosion Problems of Ammonium Sulphate Manufacture. A. W. Bamforth. (*Chemical and Process Engineering, Corrosion Technology*, 1954, **1**, Apr., 31–35). In the manufacture of ammonium sulphate, the erosive effect of crystals present in solution, the low sulphuric acid concentrations, and the relatively high temperatures of operation give rise to corrosion problems. These are discussed in detail and advice is given on materials of construction. (50 references).—L. E. W.

Effect of Low-Temperature Stress-Relieving on Stress-Corrosion Cracking. C. R. McKinsey. (*Welding J.*, 1954, **33**, Apr., 161s–166s). Tests were made under laboratory conditions to show that residual welding stresses can be sufficiently high to cause stress-corrosion cracking in some steels when exposed to a mixed nitrate solution. Both rimmed and killed steels were susceptible, and the carbon content appeared to have some effect. Controlled low-temperature stress-relieving protected welded plates from stress-corrosion cracking. Field tests were also carried out. Severe cracking took place in as-welded plates when immersed in 85% ammonium nitrate solution for 10 months.—U. E.

Corrosion of Nickel Cast Irons in Soils. I. A. Denison and M. Romanoff. (*Corrosion*, 1954, **10**, June, 199–204). Corrosion tests in many soils up to 14 years showed that up to 3% of nickel frequently reduced the initial corrosion of cast iron, but the improvement was not always maintained. Austenitic cast iron containing 15% Ni, 6·6% Cu, and 2·6% Cr corroded much less rapidly. Bursting tests up to 500 lb./sq. in. indicate that severely corroded pipe can withstand such pressures owing to the nature of the corrosion products.—J. F. S.

Inspection of Petroleum Refinery Equipment. E. H. Tandy. (*Corrosion*, 1954, **10**, May, 160–164). The parts of refinery equipment most liable to corrosion damage are described and inspection procedures outlined.—J. F. S.

Prevention of Corrosion by Water in Oil Refineries. P. W. Sherwood. (*Corrosion Technology*, 1954, **1**, June, 113–115, 120). The author points out that protection from corrosion by water is the most important single item of preventive maintenance in petroleum refineries. The three principal variables which determine the degree of corrosiveness, namely: (1) presence and nature of impurities in the water; (2) temperature of the system; and (3) velocity of the water; are dealt with. The author discusses the methods of control with particular reference to cathodic protection, metallic coatings, organic coatings, inhibitors, alteration of environment, and choice of metals.—L. E. W.

Water Treatment for Cooling Towers. W. J. Winzig. (*Blast Furn. Steel Plant*, 1954, **42**, July, 812–816). Methods used for the treatment of water in cooling towers to prevent corrosion, formation of scale, organic growths, and delignification of the wood packing of the towers are considered.—B. G. B.

Contamination of Nitric Acid as a Source of Error in the Huey Boiling Nitric Acid Test. R. J. Bendure. (*Corrosion*, 1954, **10**, Mar., Corrosion Control 1). Abnormally high attack rates were observed when hydrofluoric acid was evaporating under the fume hood being used for a Huey 65% boiling nitric acid test. Hydrogen sulphide is believed to interfere similarly.—J. F. S.

Research on the Acid Resistance of High Chromium Stainless Steels. (I, II). M. Tagaya and S. Isa. (*Tetsu to Hagane*, 1953, **39**, Nov., 1266–1270; Dec., 1330–1335). [In Japanese]. Corrosion tests in boiling 5% H_2SO_4 were made on 27/5 Cr–Ni, 27/5/1/1 Cr–Ni–Mo–Cu, and 27/10/1/1 Cr–Ni–Mo–Cu steels under various conditions of heat-treatment between 700° and 1100° C., formation of sigma phase, and surface treatment. The first had the lower, and the second the higher acid resistance, and surface treatments improved the resistance. The effect of reducing the chromium content of 27/5/1/1 Cr–Ni–Mo–Cu steel was investigated, and it was found that the proportion of 25% Cr was the lowest limit for acid resistance.—K. E. J.

The Mechanism of Inhibition of the Corrosion of Iron by Sodium Hydroxide Solution. Part II. J. E. O. Mayne and J. W. Menter. (*J. Chem. Soc.*, 1954, Jan., 99–103). The film formed on iron by the anodic discharge of hydroxyl ions is composed of material having the cubic structure of either Fe_3O_4 or γ-Fe_2O_3; consequently it is indistinguishable from the air-formed film. Freshly abraded iron becomes passive when immersed in 0·1*N* sodium hydroxide containing dissolved oxygen. The mechanism of this is explained.

The Corrosion of Castings and a New Concept of " Balanced *p*H " (*p*Hp). A. Levasseur. (*International Foundry Congress, Paris*, Sept. 19–27, 1953, *Paper No.* CI–6). [In French]. After briefly considering the chemical corrosion of castings the author points out that for liquids containing very little water the normally measured *p*H is not an accurate measure of the acidity or basicity, as it assumes that the concentration of undissociated H_2O molecules is the same as for pure water. Under these conditions it is suggested that the acidity or basicity may be defined as the *p*H value of an aqueous solution for which the hydrogen and hydroxyl ion concentrations are the same as for the liquid under study. The value of this " balanced *p*H " (*p*Hp) is given by $p\text{H} - \frac{1}{2}\log 55{\cdot}5/(\text{H}_2\text{O})$ where (H_2O) is the molecular concentration of water expressed in g.mol/litre.—B. C. W.

Study on High Temperature Oxidation of Heat Resisting Steels (I). T. Akutagawa and T. Fujita. (*Tetsu to Hagane*, 1954, **40**, Jan., 31–38). [In Japanese]. " Vanadium attack," arising from heavy oil burning in gas turbines, increases the loss of metal, and is attributed to the destruction of the protective MoO_3 film on the metal and to the catalysis of oxidation. It can be reduced by bringing CaO or MgO into contact with the V_2O_5, or by using a protective ceramic coating on the turbine metal.—K. E. J.

ANALYSIS

Determination of Sulfur in Iron and Steel by Barium Chloride Method after Chromatographic Separation of Sulfuric Acid. F. Nydahl. (*Analy. Chem.*, 1954, **26**, Mar., 580–584). Most of the complications in the barium chloride method, caused by interfering ions, can be eliminated if, before precipitation, the sulfuric acid is chromatographically separated from the solution. Aluminium oxide is used to adsorb the HSO_4^- ion, the selectivity being very good. Details of the method are given. The method was tested on standard steel and iron ore samples.—T. E. D.

Application of the High Voltage A.C. Arc Source to Spectrographic Analysis of Steels. T. Mitsuhashi, T. Nakashima, and Y. Shiraishi. (*Tetsu to Hagane*, 1953, **39**, Nov., 1277–1281). [In Japanese]. An apparatus using 4400 or 2200 V. arc sources is described; it can be used for spectrographic analysis of steel for manganese, titanium, silicon, nickel, chromium, and arsenic, and gives results for small amounts which are impossible using spark methods. The method has high sensitivity, and reproducibility within 3–8%; there is practically no continuous background, and little loss of the specimen.

Improvement of Melting Apparatus for the Vacuum Fusion Oxygen Analysis by the Carbon Cylindrical Furnace. Y. Shimokawa. (*Tetsu to Hagane*, 1953, **39**, Dec., 1342–1350). [In Japanese]. A cylindrical furnace was used instead of the usual induction or carbon spiral furnace; the advantages are ease of manufacture of heating units, ease of operation, and high and readily controlled temperature. Results of some determinations and comparisons with other methods are given; results agree with those from the Herty method within 10%. (16 references).—K. E. J.

A Statistical Study of Comparative Analyses of Cast Iron. J. Gelain. (*International Foundry Congress, Paris*, Sept. 19–27, 1953, *Paper No.* CI–3). [In French]. A statistical study has been made of determinations of carbon, manganese, silicon, sulphur, and phosphorus, obtained for the same materials by 80–95 different laboratories. It is concluded that for a 0·95 probability the accuracy between laboratories for carbon and silicon analyses is $\pm$ 5%, for manganese $\pm$ 10%, and for sulphur $\pm$ 30%. The results of the phosphorus determinations were very irregular.—B. C. W.

The Determination of Total Sulphur in Solid Fuels. E. Strambi. (*Calore*, 1954, **25**, Mar., 111–118]. [In Italian]. The author reviews present methods of determining sulphur in solid fuels and indicates that these are either too slow or insufficiently accurate. He puts forward a method which consists of burning a mixture of coal, carbonate of soda, and magnesium oxide in a calorimeter at a pressure of 25 atm. and with oxygen enrichment. The products of combustion are washed in hot water and the sulphur is determined as barium sulphate. The author claims simplicity, economy, speed, and reliability for his method. (18 references).

ECONOMICS AND STATISTICS

Iron and Steel Production in Central Africa: Present Position and Future Prospects. (*Iron Coal Trades Rev.*, 1954, **169**, July 30, 300–302). An account is given of the present iron and steel producing capacity of Central Africa, and its sources of cheap and plentiful iron ore and limestone are indicated. It is concluded that the area shows great promise for further expansion.—G. F.

The Transport of Fuel and Minerals Used in the Lorraine Iron and Steel Industry—Tariffs on Iron Ores and Fuels. (*Ann. Mines*, 1954, June, 12–21). A detailed review of the cost of transporting iron ore and fuel in this region, is made.

The Belgian Iron and Steel Industry. L. Bekaert. (*Eastern Metals Rev.*, 1953, **6**, Feb. 2, 121–124, 155). The organization and work of the Belgian iron and steel industry are described and the activities of Fabrimetal (the trade association of 2000 works) is discussed.—B. G. B.

India's First Five-Year Plan and Iron and Steel Industry. R. N. Kapur. (*Eastern Metals Rev.*, 1953, **6**, Feb. 2, 107–109). A short account is given of recent developments in the Indian iron and steel industry.—B. G. B.

MISCELLANEOUS

Some Facts on the Activities of L'Institut de Recherches de la Sidérurgie (IRSID). M. Allard. (*Usine Nouvelle*, 1954, Spring Issue, 175–179). The French iron and steel research organization came into being in 1947 and now has a staff of about 100 engineers and 300 technicians and assistants, many of whom were recruited from the industry. Work in which IRSID participates is very varied, ranging from investigations on the blast-furnace, low-shaft furnace, and basic-Bessemer converter, to physical and metallurgical studies in the laboratory.—T. E. D.

Physics and the Iron and Steel Industry. P. Bastien. (*Usine Nouvelle*, 1954, Spring Issue, 5–15). A full summary of the various applications of physics in the iron and steel industry is given in three sections: (*a*) Methods of control during manufacture; (*b*) techniques for use in laboratory investigations; and (*c*) the physics of metals applied to the metallurgy of iron and steel.—T. E. D.

The Structure and Working Methods of a Centralized Materials Management Department. K. Lehmann, H. Scholten, and O. Wagner. (*Stahl u. Eisen*, 1954, **74**, Aug. 26, 1113–1128). The basis for the introduction of a central materials management in a steelworks, its widening field of activity, and its present organization are discussed.—J. P.

Maintenance Policies and Methods. L. A. Ferney. (*Engineer*, 1954, **198**, Aug. 6, 185–186). The author stresses the need for carrying out maintenance with certain definite aims in view. These aims are mentioned and the procedures of maintenance are examined in relation to them. It is shown that for any given plant there should be a defined minimum cost associated with a particular degree of preventive maintenance.—M. D. J. B.

Boiler Water Specifications. T. H. Turner. (*Engineer*, 1954, **197**, June 25, 918–919). The first stage of British Standards Institution work undertaken at the request of Lloyd's Register in 1942 has been completed by the publication of two further specifications. Wide use of these is recommended to all boiler users.—M. J. D. B.

The Use of Oxygen in the Metal Industry—Tasks and Aims. L. Jeníček. (*Hutník*, (Prague), 1954, **4**, (6), 182–184). [In Czech]. A report is given on the principal contributions to the Conference on the Utilisation of Oxygen in the Metal Industry, and its Production, held at Liblice under the auspices of the Czechoslovak Academy of Sciences in the Spring of 1954.—P. F.

Works Transport. (*Engineering*, 1954, **178**, Aug. 27, 273–277; Sept. 3, 300–301). An account is given of a two-day conference on works transport organised by the British Iron and Steel Research Association. The conference covered the organization of works traffic, congestion problems, the economic aspects of transport, maintenance of locomotives, the relative merits of steam and Diesel traction, and wagons for bulk commodities.—M. D. J. B.

Measuring with Electrons. (*Steel*, 1954, **134**, May 24, 104–106). Applications of load cells to measure weight and force are described. They can be made to operate automatic controls and alarms. Examples cited include: Weighing a 32-ton steel casting; proportioning and weighing moulding sand; and controls for rolling mills.—D. L. C. P.

Closed-Circuit Television for Industry. L. Walter. (*Iron Coal Trades Rev.*, 1954, **169**, July 30, 282–284). The author describes some of the new industrial closed-circuit television equipment now available and briefly reviews some of its present and possible future uses in the iron, steel, and engineering industries, and in research and education.—G. F.

Social Organization of Work in France. P. Cornet. (*IRSID, Centre d'Etudes supérieures de la Sidérurgie, Metz*, 1952, "Les Cahiers du CESSID," No. VI). The evolution of the present system is described and consideration is given to the regulation of working hours and its application in the metallurgical industries.—A. G.

Aluminized Safety Clothing Keeps Cool in the Hot Jobs. (*Steel*, 1954, **134**, June 14, 118–121). The use of a new lightweight aluminized cloth for protective clothing is reported.

Air Pollution—A Bibliography. S. J. Davenport and G. G. Morgis. (*U.S. Bureau of Mines, Bull.* 537, 1954).

BOOK NOTICES

HAMPEL, C. A. (Editor). "*Rare Metals Handbook*." La. 8vo, pp. xii + 656. New York, 1954: Reinhold Publishing Corp. (Price 96*s*.). The author has tried to choose a title for this volume of studies of the less widely-used metals which will not be misleading, but the one he has hit upon does not, in fact, really unite its 35 or so subjects. They range from silicon, present in the earth's crust at 277,200 parts per million, to ruthenium, at 0·001; they do not, however, include copper, at 70 parts per million, nor nickel at 80, but they do include manganese at 1000, and titanium at 4400. All the same, there is to be found here a great deal of valuable information on what may be called the quasi-familiar metals, such as germanium, gallum, and selenium, and the scarcely-familiar, such as hafnium and rhenium. The book will form a useful collection of information, though it provides little that is new or original; except perhaps in the chapter on uranium, which is far the longest and contains references as recent as 1952, the sources of material tend to be old and well-tried and available elsewhere. The volume's merit lies in the breadth and compactness of the contents, not in any new approach or unpublished information.—J. P. S.

SÉFÉRIAN, D. "*Les Soudures. Techniques—Controle—Soudabilité des Métaux*." Deuxième edition. Préface de P. Chevenard. 8vo, pp. xx + 428. Illustrated. Paris, 1954: Dunod. (Price 3600 fr.).

The present is the second edition of a book originally published in 1948. It seems that the author has endeavoured to bring it up to date by making additions rather than by modifying the existing textual setting where required—hence the scanty mention of welding in an inert atmosphere, with no reference to the use of that process with consumable electrodes in the chapter on the welding of magnesium, and the addition of stud welding outside the chapter on metal-arc welding to which it rightly belongs.

The book is divided into four parts. The first deals with oxy-acetylene and metal arc welding, oxy-acetylene welding flames and welding methods, arc welding electrodes and methods employed, and automatic operation. Highly theoretical calculations, the value of which may appear doubtful because of the difficulty of covering various practical conditions, are also given. The second part contains chapters on the testing and inspecting of welds and on the defects which can occur. The third part (the longest section, occupying 147 pages) is concerned with the weldability of materials, special mention being made of carbon and special steels, cast iron, aluminium, copper, nickel, zinc, magnesium and their alloys, and also lead. The subject is introduced by chapters on the phenomena occurring in the weld and base metal during welding and on weldability tests. The last part covers other welding processes, and related techniques. Resistance welding, hard-facing, oxygen-cutting, brazing, stud welding, and the welding of plastics are each given a chapter. The book ends with a bibliographical list of references covering five pages.

The book is written in an admirably clear style and is well illustrated. Despite its little weaknesses indicated above, it should be most useful to all those wishing to study the practical and theoretical aspects of welding.

F. KOENIGSBERGER.

AMERICAN IRON AND STEEL INSTITUTE. "*Survey of Furnace Availability and Roof Life.*" (Contributions to Metallurgy, No. 45). La. 4to, pp. 266. Illustrated. New York, 1954: The Institute.

AMERICAN SOCIETY FOR TESTING MATERIALS. "*Symposium on Effect of Temperature on the Brittle Behaviour of Metals with Particular Reference to Low Temperatures.*" (ASTM Special Technical Publication No. 158, presented at the Fifty-Sixth Annual Meeting, American Society for Testing Materials, Atlantic City, N.J., June 28–30, 1953). 8vo, pp. vi + 474. Illustrated. Philadelphia, Pa., 1954: The Society. (Price $7.50)

AMERICAN SOCIETY FOR TESTING MATERIALS. "*Symposium on Radioactivity—An Introduction. (Radioisotopes—Laboratories—Personnel—Radiation—Management Problems—ASTM Work).*" (ASTM Special Technical Publication No. 159, presented at the Fifty-Sixth Annual Meeting, American Society for Testing Materials, Atlantic City, N.J., June 30, 1953). 8vo, pp. [iv +] 46. Illustrated. Philadelphia, Pa., 1954: The Society. (Price $1.40)

"*Atlas zur Wärmebehandlung der Stähle.*" Herausgegeben vom Max-Planck-Institut für Eisenforschung in Zusammenarbeit mit dem Werkstoffausschuss des Vereins Deutscher Ingenieure. Teil I, von Franz Wever und Adolf Rose. Teil II, von Adolf Rose, Walter Peter und Werner Strassburg. La. 8vo, pp. 147, loose-leaf. Illustrated. Düsseldorf, 1954: Verlag Stahleisen, m.b.H. (Price DM 70.–. Personal members of Verein Deutscher Eisenhüttenleute. DM 63.–)

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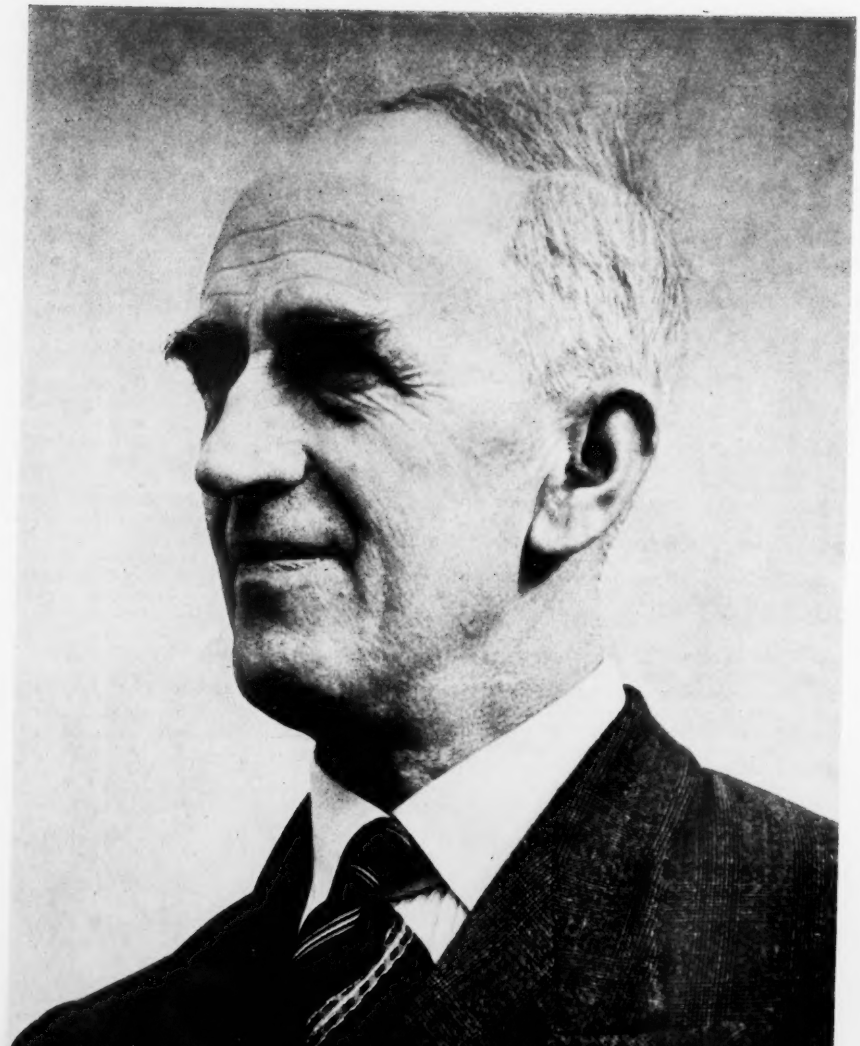
H. D. WARD

HENRY D. WARD was born in 1899, and was educated in the Isle of Man. During the First World War he served in the Tank Corps.

Mr. Ward first entered commerce when he joined the Monk Bridge Iron and Steel Company, Limited, and he remained with that Company until three years ago. During this period he was concerned with steelmelting, first as a Chemist and later as Open-Hearth Shop Manager. For the last three years he has been engaged in the manufacture of jet aircraft blanks with Daniel Doncaster and Sons, Limited.

Mr. Ward has always been closely connected with metallurgical education in Leeds, and for several years he was a part-time lecturer in Metallurgy at the Leeds College of Technology. He has served on the Committee of the Leeds Metallurgical Society since its inception in 1947, and was elected President for the Session 1953–54.



H. D. Ward

Honorary Member of Council, 1953–54